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Non-Ferrous Trade Developments

By Gilbert Evans

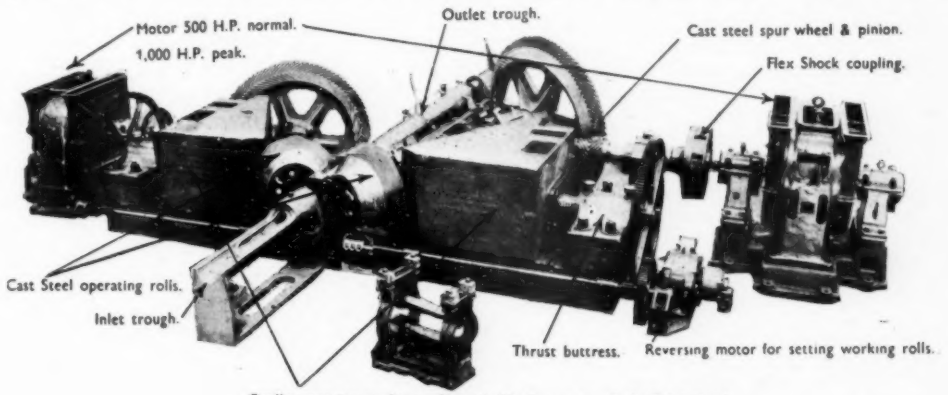
Tube manufacturing conditions in Germany are briefly reviewed and comparisons are made with the production methods used in Great Britain. Rotary piercing methods are discussed and a comparison is made with the extrusion process; the fields of usefulness of each process are also briefly considered.

ONE of the most important developments in the metal trades is the introduction of the four-high high-speed rolling mill by Messrs. W. H. A. Robertson and Co. Ltd. This design of mill is capable of dealing with the production of plates, strips and the like at speeds up to 500 ft. per min., and it has proved to be a great advancement on the long-established three-high mill. A large number of mills are in course of construction, while others have been installed in many British plants and are now contributing to the speed-up in production. The interests of the writer, however, are chiefly centred in the manufacture of tubes, and he was for thirteen weeks, just before the outbreak of the present war, investigating tube manufacturing conditions in Düsseldorf, Remschild, Mannheim, Witten, and several places in Germany, being fortunate to return to England five weeks before war was declared.

Having visited German tube manufacturing plants on many occasions during the past thirty years, several times in the interests of British manufacturers, the writer was surprised to find little of note in his recent visit with regard to rotary methods, in comparison with those in operation in Great Britain. The disc-piercing mill was preferred to the three rolls method for sizes up to 6 in. diameter, and the latter process for the production of tubes up to 16 in. outside diameter by ¼ in. thickness of walls at a speed which was slow and cumbersome in comparison with some of the steel tube mills operating in this country. On the other hand, it was surprising to find that in the production of non-ferrous tubes, mainly copper alloys, the disc mill—namely, the Mannesmann and Steifel types—was still in general use, a method which, with regard to quality and quantity of production, is not comparable with the Evans' rotary piercer.

To obtain a medium-sized tube of, say, 4 in. diameter, it was necessary to use a 5-in. solid billet and press it twice through the rolls, first piercing and then expanding; whereas by the Evans' method the same size of tube can be obtained from a 4-in. solid billet by one operation in the piercing mill. The time taken with the latter method is 2½ secs., compared with 2 mins. for the piercing and expanding processes in German practice. It is of interest to note that a complete tube-making plant was installed in the works of a Scandinavian firm by W. H. A. Robertson and Co. in the early days of the war. This modern mill, with a flexible capacity in diameters up to 18 in. by ½ in. thickness of wall and lengths of 12 ft., was shortly afterwards taken over by the Germans. In operation the output of this mill was about 16 tons per 8-hour shift.

Since that time several Evans' rotary mills have been put into operation in Great Britain, the largest of which is shown in the accompanying illustration. This mill is of rather an outstanding character and some brief information regarding it will be of interest. The complete mill weighs 38 tons, and it is driven by variable-speed motors which ensure a peripheral speed suitable for different sizes of tube and for different alloys to be processed. Its flexibility is an important feature, since it is capable of producing tubes from 2 in. to 18 in. inside diameter, possessing any desired wall thickness. The average length of the billet used is 6 ft., and the average length of the pierced shell 10 ft. 6 in. to 12 ft. The rolls revolve at a speed of 75 per min., and the time of operation from 25 to 30 secs. The number of operators varies with the weight of billet being processed, and may be from 5 to 7. The adjustments necessary for producing different diameters are quickly made, and many different diameters of tube can be produced in any one shift. Some idea of the production from this machine can be gathered from the following output table from different sized billets :—


Motor 500 H.P. normal.
1,000 H.P. peak.
Outlet trough.
Cast steel spur wheel & pinion.
Flex Shock coupling.
Cast Steel operating rolls.
Inlet trough.
Thrust buttress.
Reversing motor for setting working rolls.
Cradles carrying working rolls with 8" adjustment. Right & left hand.

By courtesy of W. H. A. Robertson and Co., Ltd.

Very large piercing mill for seamless copper tubes now in operation in Great Britain. Designed on Evans' patent, it will take billet from 2 in. to 11 in. in diameter and will expand up to 18 in. inside diameter to a desired wall thickness.

2 in.	diameter billets	10 tons per 9-hour shift.
3 in.	"	14 tons "
4 in.	"	16 tons "
5 in.	"	18 tons "
6 in.	"	22 tons "
7 in.	"	26 tons "
9 in.	"	30 tons "

In 1914 the writer was supervising the production of copper and copper alloy tubes, varying in diameters from $\frac{5}{8}$ in. to 12 in., by means of the rotary piercing process, and casting in cores for subsequent finishing by cold-drawing—producing upwards of 84 tons per week. With this fact in view, it is surprising that technical advisers to the Ministry of Supply have stated that the modern problem of shell bands has been simplified by the adoption of cupping and drawing from sheet and by means of sheet instead of rings. The scrap produced by a process rejected as obsolete in 1901-1902 must be at least 25%, compared to a production efficiency of about 95% by rotary piercing methods.

Reverting to the writer's recent investigations in Germany, the one method in which advancement had been made was in the extrusion of low-carbon steel tubes by the Erhardt extrusion rolling and reducing methods. One-inch tubes up to 40 ft. long and within limits of standard requirements, were produced at the rate of 80 per hour. The extruding presses were of the mechanised crank type, and the finished tube was of excellent quality and finish. The whole process consisted of reheating the billet, which was passed through a descaling mill of the reeler type, on to the extrusion press and then direct to a Diescher reducing mill, on to sizing rolls, reeler straightened, and finally to the cooling rack, after which it was cut to finished lengths and delivered to storage. The whole plant was mechanically operated, even to the initial charging of the furnace.

Several complete plants for the extrusion of copper and its alloys have been built by Messrs. Fielding and Platt, Ltd., and installed in British factories since the outbreak of war. One of these, with which the writer was in contact, is of horizontal design and operates at a pressure of 1,500 tons; it requires a leading hand and three assistants to operate the plant, which gives excellent results on some of the delicate alloys, but the cost in mandrells and dies is high. Difficulties have been experienced in obtaining the grade of steel possessing sufficient toughness for these purposes.

A question frequently arises regarding the finish of the tube, as sent to the drawing mill, from the extrusion press, in comparison with the shell produced by the rotary piercer. The writer is of the opinion that the result is favourable to the rotary produced shell, this product being free from scratches due to die trouble and mandrell failure, especially in the smaller diameters. It should be necessary to change tools, mandrell and die, after each operation, but this plan of working on a relay of tools, so giving each set time to cool before being again brought into use, is possible only in the case of the matrix or die. Different conditions apply to the Fielding and Platt and the Serck vertical presses, both of which are constructed by Fielding and Platt, Ltd.

It is necessary to state that, for small tubes up to 2 in. in diameter the writer prefers the vertical press to the rotary process, although the shortness of the billet confines the weight to a marked degree. On the other hand, the rotary process is best for tubes over 3 in. diameter, and it is emphasised that no tube plant is complete without at least one unit of each type. It can be stated that the extrusion method is superior in its scope of application for copper-base alloys, while the piercing mill occupies pride of place in its application to copper itself, and a few alloys, such as cupro nickel, arsenical copper, and the like.

In view of the great advance made in plants for these products, it is indeed strange to find some small producers of tubes adhering to the old-fashioned method of casting in cores in split moulds. The term tube finishers could properly be applied to them, rather than tube manufacturers; such firms are sometimes able to obtain hollow shells from firms which have adopted the rotary mill, but as often as not are bound to take a pierced shell out of all proportion to the finished article, involving increased cost of manufacture due to the greater number of picklings, annealing and cold-drawing operations involved.

A great deal of controversy has arisen from time to time over that bugbear of all tube makers, concentricity. In spite of all the new gadgets, patented and otherwise, such drawbacks as the sweating of the material in the furnace has to be considered. All kinds of heating mediums have been experimented with and are used, but the ideal heating medium or conditions has not yet been discovered.

A case occurred in which alloy billets containing spelter were left, at the end of the working shift, in the heating furnace overnight and found to have lost appreciably in diameter by the next morning, and in some instances spelter had volatilised during this soaking time. In another case long copper billets for rotary piercing, similarly treated in the furnace, had flakes over $\frac{1}{16}$ in. thick, due to oxidation of their surfaces. It was no rare event for material so treated to break into short pieces on being drawn on to the foreplate of the furnace, or even to collapse, when gripped by the tongs for transfer from furnace to piercing unit, and most certainly on the billet making contact with the working planes of the mill. It should be an unbroken maxim of those responsible for production to insist that all furnaces are discharged before ceasing operations, and to ensure that billets are brought to the required temperature with the increasing temperature of the furnace on each shift. Such care is not so necessary when night and day shifts follow each other.

In conclusion, the writer wishes to take this opportunity of paying tribute to the late Mr. Kenneth Fraser, with whom he was associated in the reorganisation of the Yorkshire Copper Works, Ltd. The association, which lasted from 1907 to 1914, proved to be the happiest one spent during 30 years in the trade. Mr. Fraser was an ideal and untiring organiser; he was terse and taciturn to a degree, and tolerated no slackness, being satisfied only with the best. After his promotion to manager and subsequently to managing director, he proved a model of efficiency, and the results of his efforts are reflected in the present plant of the Yorkshire Copper Works, Ltd., which is probably one of the most modern of its kind in the world.

The Spectrographic Analysis of Tin-Lead Solders

AN investigation carried out by the British Non-Ferrous Metals Research Association and reported by D. M. Smith* forms an extension of the work previously reported† on the spectrographic analysis of tin, and deals with the quantitative determination of small amounts, of the order of 0·1% and less, of aluminium, bismuth, cadmium, copper and zinc in tin-lead solders.

In view of the difficulty in obtaining satisfactory chemical checks on the composition of these alloys, synthetic residues have been prepared by mixing solutions of the various constituents in the appropriate proportions and evaporating to dryness. Residues of alloys, obtained by dissolving and evaporating, have been compared with the synthetic residues by means of the graphic arc. The method was found to be useful in obtaining a series of standard alloys that can be used with confidence, but is too tedious for use in routine analysis.

Analytical tables based on the "internal standard" method have been compiled from spark spectra obtained by the use of metallic electrodes following the usual procedure. It has been observed that the solder spectrum resembles the tin spectrum in many respects, but shows marked differences as compared with the lead spectrum. Conditions of spark excitation standardised for the analysis of tin have therefore been adopted.

In the case of the determination of bismuth, the presence of 2·5 to 2·8% antimony appears to have no effect on the relative intensities of the lines in the solder spark spectrum, as compared with that of an antimony-free solder.

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* Publication No. 105, issued by the Tin Research Institute, Fraser Road, Greenford, Middlesex.

† Brit. Non-Ferrous Metals Res. Assoc., R.R.A. No. 427, Sept., 1936 (I.T.R.D.C. Tech. Publ. A., No. 16, Nov., 1936) and R.R.A. No. 515, June, 1939.

The Constitution of Alloys of Aluminium with Magnesium and Iron

By H. W. L. Phillips, M.A.

The constitution of alloys of aluminium with 0–5% magnesium and 0–2·5% iron has been studied in detail, and the position of the binary valley bounding the primary aluminium phase field has been redetermined. Little was known of the constitution of these alloys as ordinarily prepared, but it was expected that metastable structures, due to undercooling and segregation, would be liable to occur. This view has been substantiated; the constituent $FeAl_3$ is easily undercooled, and free Mg_2Al_3 makes its appearance at percentages of magnesium well below the equilibrium solubility. The report, of which this is an abridgement, is one of a series presented to the Institute of Metals.*

THIS investigation is one of a series on the constitution of aluminium alloys under conditions of metastable equilibrium, such as are likely to be met with in commercial practice. Previous investigations have dealt with the constitution of alloys of aluminium with iron and silicon,[1] with these elements and copper,[2] and with magnesium and silicon.[3] The work has been carried out in the Research Laboratories of the British Aluminium Co., Ltd., under the general supervision of Dr. A. G. C. Gwyer, B.Sc.

Previous studies of the constitution of aluminium with magnesium and iron are reviewed. According to Fuss[4] and Barnick and Hanemann,[5] the section $FeAl_3$–Mg_2Al_3 is quasi-binary, and the partial system the components of which are aluminium, $FeAl_3$ and Mg_2Al_3, is eutectiferous. Fuss considers that the ternary eutectic of these three phases is almost coincident with the binary eutectic of aluminium and Mg_2Al_3, and contains about 35% magnesium and 65% aluminium with hardly any iron, freezing at 448° C. Barnick and Hanemann, however, place the ternary eutectic at 29% magnesium, 3% iron, 68% aluminium, and the freezing point at 445° C. Both diagrams have been plotted in Fig. 1.

The present investigation forms a necessary preliminary to the study of the quaternary alloys of aluminium with magnesium, silicon and iron. In view of the discrepancy between the work of Fuss and that of Barnick and Hanemann, it was impossible to use their data, and both diagrams are only of use in tracing the changes taking place below the liquidus if it is assumed that equilibrium is maintained—a condition which previous experience of aluminium alloys has shown to be one that is not likely to be realised.

It was decided to reinvestigate the course of the binary valley between the aluminium and $FeAl_3$ primary fields from the binary eutectic to the ternary eutectic, and to study in detail the range 0–2·5% iron, 0–5% magnesium. The experimental methods and materials used were identical with those described in the author's previous paper.[3] Iron was added in the form of a hardener, made up from Armco iron and super-purity aluminium, a typical analysis being:

Iron	...	...	...	10·28 per cent.
Silicon	...	...	...	0·01 " "
Copper	...	...	...	< 0·01 " "
Manganese	...	...	...	< 0·01 " "

Micrography of the Alloys

Using super-purity aluminium as a basis, the only constituents found to occur were $FeAl_3$,† Mg_2Al_3, and the aluminium-rich solid solution forming the continuous phase. $FeAl_3$ crystallises in needles or prisms, of lavender grey colour, and presents no difficulty in identification. Mg_2Al_3 is almost impossible to distinguish from aluminium without etching, but is outlined and slightly darkened by a momentary dip in dilute sodium hydroxide solution.

Using aluminium of ordinary purity as a basis, other constituents, due to silicon, make their appearance. Of these, Mg_2Si is the most widely prevalent; it occurs as

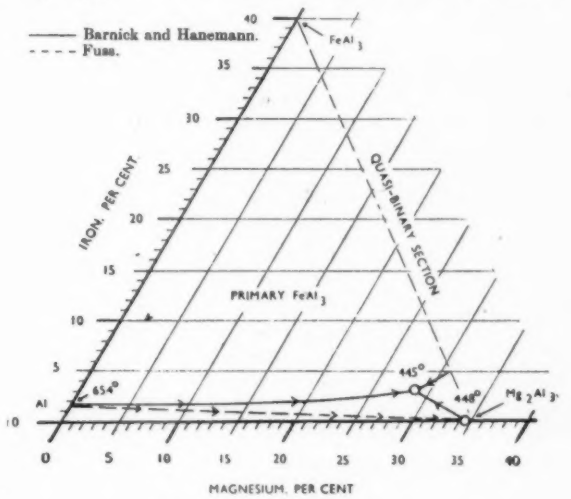

Fig. 1.—General survey.

isolated particles or minute clusters in "script" formation, and can be identified by its pale blue colour and ease of attack by etching reagents. In alloys containing low percentages of magnesium, the aluminium-iron-silicon constituents α(Fe-Si) and β(Fe-Si) also occur. The former,

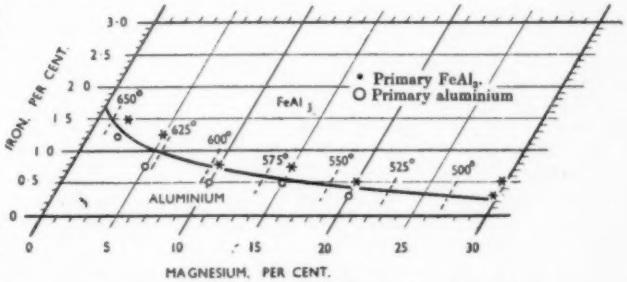

Fig. 2.—The aluminium-$FeAl_3$ binary valley.

like Mg_2Si, occurs in script form, but is darker, more in relief, and differs from it in etching behaviour. β(Fe-Si) occurs in the form of needles.

General Survey

Over the range studied, two primary phase fields occur, due to the aluminium-rich solid solution and $FeAl_3$. These

* *J. Inst. Metals*, 1941, **67**, 275.

† For convenience, the name $FeAl_3$ is retained, although Bradley and his colleagues have shown that the constituent decomposes into Fe_2Al_5 and Fe_2Al_7 on slow cooling to room temperature. Cf. *J. Inst. Metals*, 1939, **65**, 410.

1 A. G. C. Gwyer and H. W. L. Phillips, *J. Inst. Metals*, 1927, **38**, 29.
2 A. G. C. Gwyer and H. W. L. Phillips and L. Mann, *J. Inst. Metals*, 1928, **40**, 297.
3 H. W. L. Phillips, *J. Inst. Metals*, **67**, p. 257.
4 V. Fuss, "Metallographie des Aluminiums und seiner Legierungen," Berlin, 1934, p. 141.
5 M. Barnick and H. Hanemann, *Aluminium*, 1938, **20**, 533.

fields are separated by a binary eutectic valley, the course of which is shown in Fig. 2. It terminates in the ternary eutectic, with freezing point 451° C. It was impossible to determine accurately the composition of the ternary eutectic, but the available evidence supports the view of Fuss, that it contains very little iron. The binary valley points lie approximately on a straight line which, when produced, cuts the eutectic horizontal (655° C.) at 1·7% iron—a figure agreeing exactly with that reported by Dix. Attempts to determine the eutectic composition by analysing eutectic areas of microsections of slowly cooled samples led to values between 1·94 and 2·04% of iron, the

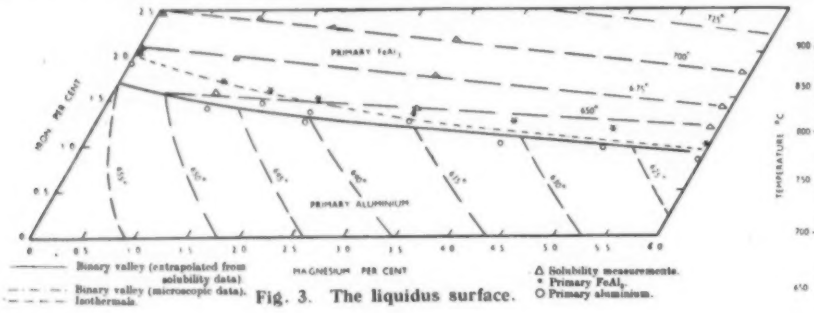

Fig. 3. The liquidus surface.

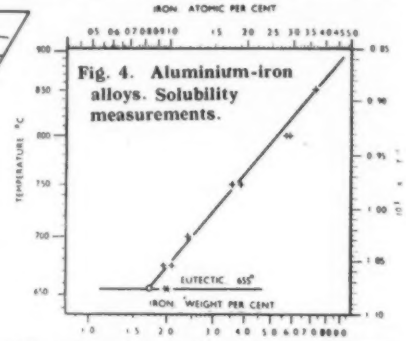

Fig. 4. Aluminium-iron alloys. Solubility measurements.

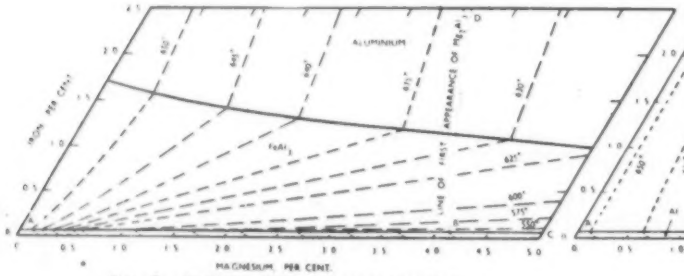

Fig. 5. Surface of secondary separation.

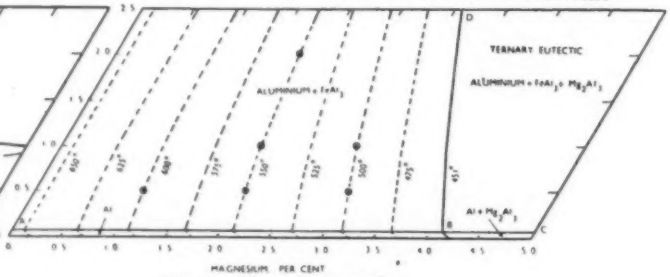

Fig. 6. The solidus surface.

is not parallel to the aluminium-mangesium axis, as with 30% magnesium the primary $FeAl_3$ field is entered with as little as 0·25% iron, giving primary crystals of the habit shown in Fig. 10.

The Liquidus Surface (Fig. 3).—In the primary aluminium field, the temperatures were obtained from cooling curves, but in the primary $FeAl_3$ field arrests due to the primary phase could not be detected over the range plotted. By increasing the iron content to about 4%, feeble arrests occurred, but owing to undercooling the temperatures were inconsistent and generally low. To obtain the isothermals of the $FeAl_3$ field, recourse was had to solubility determinations. The alloy was held molten at a constant temperature, and samples of the liquid portion were taken after one and after two hours and analysed for magnesium and iron, the former being determined gravimetrically as pyrophosphate and the latter by titration against titanous chloride. These figures were also used to confirm the position of the binary valley. Fink and Freche[8] have shown that for many binary aluminium alloys the solid solubility and hypereutectic liquidus curves can be represented by equations of the form :

$$\log_e x' = -\frac{L}{RT} + C$$

where x' = molecular fraction of alloying element, L = molecular heat of solution, R = gas constant, T = temperature, degrees absolute, C = integration constant.

This equation applies when the precipitating phase is a pure element or compound, and is based on the assumption that the solutions are perfect and that the molecular heat of solution is constant. Whilst neither of these assumptions is justifiable, Fink and Freche have shown that, in a large number of instances, the equation leads to the expected linear relationship between the logarithm of the composition and the reciprocal of the absolute temperature of precipitation.

The solubility diagram for the aluminium-iron alloys, plotted in this manner, is shown in Fig. 4. The experimental mean figure of 2·0% being indicated on the diagram by an asterisk.

The position of the binary valley in the ternary alloys was determined by finding the lowest percentage of iron at which $FeAl_3$ became primary in the microsections, and by extrapolation of the logarithmic liquidus curves to the point of intersection with the surface of aluminium separation. The values differed slightly, that obtained by extrapolation from solubility data being the lower by about 0·1–0·2%, thus confirming the prevalence of undercooling.

The Surface of Secondary Separation (Fig. 5).—Beneath the primary $FeAl_3$ field, the secondary separation is that of the aluminium-$FeAl_3$ binary complex. Well-defined arrests were observed, and the isothermals were rectilinear, radiating, as would be expected, from the point representing the $FeAl_3$ composition. Beneath the primary aluminium field, the second constituent to separate is $FeAl_3$. Even with relatively high percentages of iron, the $FeAl_3$ arrests were ill-defined and somewhat irregular ; with low percentages they were impossible to detect. Near the aluminium-magnesium axis, therefore, the secondary isothermals have been derived from theoretical considerations.

In Fig. 5 the line ABC is the line of first appearance of $FeAl_3$. Although for the sake of clarity it has been shown separated from the aluminium-magnesium axis, actually it must lie extremely close to it, since in slowly cooled alloys $FeAl_3$ can be identified without difficulty at as little as 0·01% iron, and even under equilibrium conditions it is probable[6] that 0·02% represents the maximum solid solubility. At A, the temperature of separation of $FeAl_3$ must be that of the binary eutectic ; at B, which lies on the line of first appearance of Mg_2Al_3, it must be that of the ternary eutectic. Experimental determination of the temperatures between A and B proved to be impossible, and the figures plotted have been based on the assumption that between these points the temperature falls linearly. This assumption appears reasonable, since over the corresponding range the solidus of the binary aluminium-magnesium alloys is also rectilinear. Smooth isothermals

6 A. Roth, *Z. Metallkunde*, 1939, **31**, 299.
7 E. H. Dix, Jr., *Metals Handbook* (Amer. Soc. Metals), 1939, p. 1224.
8 W. L. Fink and H. R. Freche, *Trans. Amer. Inst. Min. Met. Eng.*, 1934, **111**, 304.

were then drawn through the points along AB, and through such arrest points as were available, to the well-established points on the binary valley.

Near the aluminium corner, the isothermals appear to be curved, and suggest that the aluminium-rich phase takes an increasing amount of magnesium into solid solution as solidification proceeds. With higher percentages of magnesium the isothermals appear to become rectilinear.

The Solidus Surface (Fig. 6).—In a number of instances, solidus temperatures were determined by the heating-quench method described previously.[3] These points have been indicated on the diagram. BD is the line of first appearance of Mg_2Al_3 and to the right of this line the solidus coincides with the separation of the ternary eutectic.

Constituents Present in the Solid State.—It has not been thought necessary to include a separate diagram showing the range of occurrence of the various constituents, since this information can be derived from the solidus diagram (Fig. 6). Two micrographs have been included to show the habit of the constituents. The structure of an alloy containing 1·0% magnesium, 0·5% iron is shown in Fig. 11, and that of an alloy containing 4·0% magnesium, 1·0% iron in Fig. 12. In both sections the aluminium-rich solid solution is primary and $FeAl_3$ is secondary. In Fig. 12 some Mg_2Al_3 is also visible. As in the majority of aluminium alloys, there is little apparent association between the three constituents of the ternary eutectic, and each seems to have crystallised independently.

Secional Diagrams (Fig. 7–9).—Three typical diagrams have been included to show the general trend of the various surfaces of crystallisation.

Summary

The constitution of alloys of aluminium with 0–5% magnesium and 0–2·5% iron has been studied in detail, and the position and temperature of the binary valley bounding the primary aluminium phase field has been redetermined. The alloys were studied as slowly cooled, and no attempt was made to investigate their structures under conditions of equilibrium.

The three phases aluminium, $FeAl_3$, Mg_2Al_3 form a eutectiferous system, the ternary eutectic freezing at 451° C. Its exact composition has not been determined, but it probably contains about 33% magnesium and 67% aluminium with very little iron. The binary valley running from the aluminium-$FeAl_3$ binary eutectic to the ternary eutectic approaches the aluminium-magnesium axis as the magnesium content increases, and at 20–30% of magnesium the primary $FeAl_3$ field is entered at about 0·25% of iron. Solubility measurements were used in fixing the position of the valley and showed that, even with slow cooling, undercooling of the $FeAl_3$ was very liable to occur. The ternary eutectic of aluminium, $FeAl_3$ and Mg_2Al_3, occurred in all alloys containing more than about 4% of magnesium, a figure much below the equilibrium solubility of magnesium in aluminium, and towards the aluminium corner there was some evidence of coring of the aluminium-rich solid solution. The solidus of those alloys the final solidification of which did not occur at the ternary eutectic was determined by quenching experiments, and found to fall smoothly.

The solid solubility of iron in aluminium is less than 0·01% in slowly cooled alloys, and does not appear to be increased by the presence of magnesium.

Fig. 7. Alloys containing 1·0 per cent. iron.

Fig. 8. Alloys containing 1·0 per cent. magnesium.

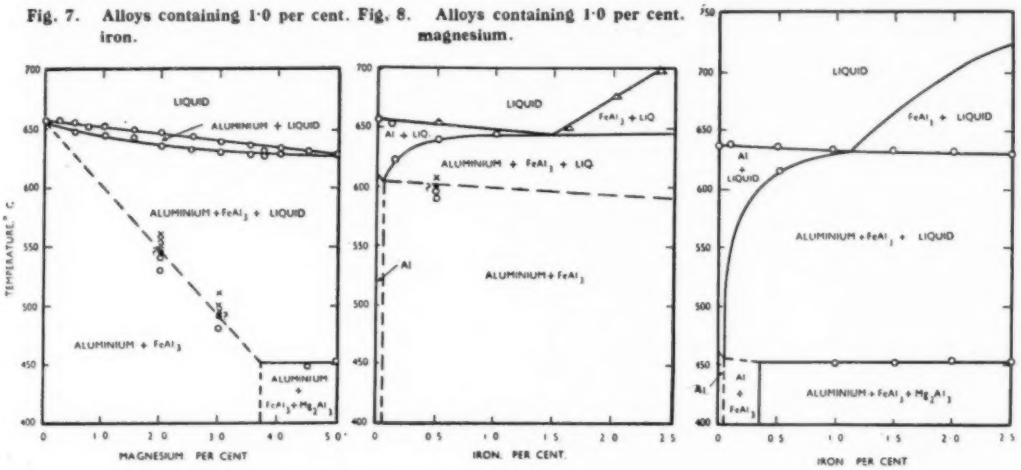

Fig. 9.—Alloys containing 4% magnesium.

Fig. 10.—Magnesium, 30.0%; iron, 0.25%; silicon, 0.020%. Etched in 12½% sodium hydroxide. x 200.

Fig. 11.—Magnesium, 1.0 %; iron, 0.5%; silicon, 0.0035%. Unetched. x 300. ($FeAl_3$: half tone.)

Fig. 12.—Magnesium 4.0%, iron 1.0%, silicon 0.0025%. Etched in 12½% sodium hydroxide. x 300.

Intergranular Corrosion in Austenitic Stainless Steels—Part IV.

By J. H. G. Monypenny, F.Inst.P.

The occurrence of intergranular corrosion in austenitic chromium-nickel steels, its detection, cause and prevention and the effects it produces, are discussed. In the practical application of these steels this trouble has been largely overcome, though not entirely, and in this article the author takes stock of the present position, studies in some detail the effect of different variables, and points out where satisfactory results can be obtained in the industrial use of the steels and where further improvements are still desirable. Attention is given mainly to the 18-8 steels, as these are the most widely used of the austenitic stainless steels, but the author emphasises that intergranular corrosion is not confined to steels of this composition.

IN the original patent covering the use of titanium as a preventive for intergranular corrosion,[37] vanadium was also included as having a similar action. Vanadium is known to be a strong carbide-former—the carbide V_4C_3 requiring a V : C ratio of 5·8 : 1—and it also has a very energetic action in preventing the formation of γ-iron, the limiting solubility of vanadium in the latter being about 1·1%. The fact that vanadium carbide appears to be but slightly soluble in iron, even at the latter's melting-point, would suggest that vanadium would be an effective addition for preventing the formation of chromium carbide in austenitic stainless steels. Recent investigations by Riedrich and Hoch[38] have shown, however, that vanadium is not very effective as a carbide-former in inhibiting intergranular susceptibility, such action as it has in this direction being due to the formation of ferrite. To appraise its action in both these ways, Riedrich and Hoch tested two series of vanadium steels. In the first series, nickel content was increased to 14% as vanadium rose to 3·0%, in order to prevent the formation of a duplex structure of austenite and ferrite; in the second series, nickel was held at 8%, so that as vanadium rose from 1·15% to 2·4% the alloys became increasingly duplex. All the steels contained about 18% chromium, and their carbon content varied from 0·09% to 0·13%. The steels were tested in the form of sheet 3 mm. thick, which was softened by heating to 1,050° C. and air cooling. Samples of the softened sheet were then reheated at 500°, 600°, and 700° C., and others were welded. Both the reheated and welded samples were then tested in the acid copper sulphate reagent. The results on the first series of steels—which were fully austenitic—are given in Table I, and indicate that steel No. 6 containing 3% vanadium and with a V : C ratio of 25 : 1, was attacked quite as badly as steel No. 1, in which the ratio was 7·6 : 1. The welded samples of all the steels showed almost complete breakdown 10/15 mm. from the welds. It would appear from these results that vanadium has little effect as a carbide-former on intergranular corrosion.

Results on the second series of steels which contained a certain proportion of ferrite are given in Table II, and they show that these steels were less susceptible to intergranular corrosion than those in the first series which were fully austenitic. Even with the highest vanadium content tested, however, the tendency to develop intergranular susceptibility was still sufficiently strong that there was some breakdown in a welded sample.

It is not necessary of course to add other alloying metals to the austenitic chromium-nickel steels to produce a duplex structure of ferrite and austenite; the latter results quite as effectively from an increase in the chromium : nickel ratio. A series of such steels were examined by Scherer, Riedrich and Hoch[39] in order to observe the effect of a duplex structure so produced on susceptibility to intergranular corrosion. The austenite in some of their steels was not stable—the result of too low a nickel content—transforming to martensite on cooling from the high treatment temperature; all such steels were free from intergranular corrosion in their tests. Results on the other steels, in which the austenite was stable, are given in Table III; the tests were carried out on 3 mm. sheet previously heat-treated at 1,050° C. followed by air cooling. The softened samples were then reheated at 600° and 700° C., as indicated in the table, and tested for intergranular corrosion by boiling for 200 hours in the acid copper sulphate reagent, except in the case of badly susceptible samples which were removed after the shorter periods indicated.

These results showed that steels containing 40% or more ferrite were completely free from intergranular attack after reheating at 600° and 700° C. for the periods mentioned in Table III; their physical properties, however, were not as good as those of the austenitic steels, in particular they were lacking in ductility. Steels containing only 5% or so of ferrite suffered almost as badly as if they were completely austenitic. Steels with an intermediate content of ferrite were susceptible after heating for an hour at 600° C., but attained complete recovery in 10 hours or longer at this temperature; they were also immune after heating for 1 hour at 700° C. Further steels were then investigated to determine the minimum content of ferrite which would

TABLE I.
INTERGRANULAR CORROSION TESTS ON AUSTENITIC STEELS CONTAINING VANADIUM.

No.	C. %.	V. %.	Reheated. 1 hr., 500° C.	1 hr., 600° C.	10 hrs., 600° C.	50 hrs., 600° C.	1 hr., 700° C.	Welded Sample.
1	0·10	0·76	No attack	Complete breakdown	Complete breakdown	Complete breakdown	Strong attack	Almost complete breakdown, 10/15 mm. from the weld
2	0·11	1·04						
3	0·13	1·32						
4	0·09	2·12						
5	0·10	2·32						
6	0·12	3·01						

TABLE II.
INTERGRANULAR CORROSION TESTS ON DUPLEX STEELS CONTAINING VANADIUM.

No.	C. %.	V. %.	% Ferrite in Structure.	Reheated. 1 hr., 500° C.	1 hr., 600° C.	10 hrs., 600° C.	50 hrs., 600° C.	1 hr., 700° C.	1 hr., 700° C., 1 hr., 600° C.	Welded Sample.
7	0·10	1·15	15	None	Complete breakdown	Some attack	Slight attack	None	None	Some attack 10/15mm from the weld
8	0·10	1·82	25	"	Almost complete breakdown	None	None	"	"	
9	0·12	2·40	30	"	"	"	"	"	"	

37 British Patent 337,349 (F. Krupp, A.G.), Oct., 27/30. The date of the corresponding German Patent is June 26/'29.

38 "Effect of Vanadium on Intercrystalline Corrosion of Austenitic Chromium-Nickel Steels," G. Riedrich and G. Hoch. *Stahl u. Eisen*, 1940, vol. 60, p. 30.

39 "Einfluss eines Gehaltes an Ferrit in Austenitischen Chrom-Nickel-Stählen auf der Korrosierfall." R. Scherer, G. Riedrich and G. Hoch. *Archiv. f.d. Eisenhüttenw.*, vol. xiii, p. 53 (July, 1939).

ensure recovery after 50 hours heating at 600° C.; from the results obtained (see Table IV), Riedrich and his collaborators concluded that the ferrite content must amount to at least 10%; probably 15% would be a safer minimum value.

Incidentally, they also tested in the same manner steels containing about 0·1% carbon and up to 25% chromium and 20% nickel, which were fully austenitic after air cooling from 1,050° C. All these steels suffered badly from intergranular corrosion after heating for either 1 hour or 50 hours at 600° C.

TABLE III.

INTERGRANULAR CORROSION TESTS ON DUPLEX CHROMIUM-NICKEL STEELS.

THE REHEATED SAMPLES WERE BOILED IN THE ACID COPPER SULPHATE FOR 200 HOURS, EXCEPT THOSE WHICH WERE BADLY ATTACKED. THE LATTER WERE REMOVED AFTER THE PERIODS STATED IN THE TABLE.

No.	C, %	Cr, %	Ni, %	Structure after A.C. 1,050° C.	Reheated			
					600° C., 1 hr.	600° C., 10 hrs.	600° C., 50 hrs.	700° C., 1 hr.
1	0·09	18·6	6·63	Aust. + 5% Ferrite	After 30 hrs.	After 100 hrs.	After 200 hrs.	After 150 hrs.
2	0·10	21·1	5·85	,, + 30% ,,	,, 200 hrs.	No attack	No attack	No attack
3	0·09	20·3	8·33	,, + 5% ,,	,, 30 hrs.	After 100 hrs.	After 200 hrs.	After 150 hrs.
4	0·09	22·2	6·05	,, + 35% ,,	,, 200 hrs.	No attack	No attack	No attack
5	0·10	22·3	8·37	,, + 20% ,,	,, 200 hrs.	,,	,,	,,
6	0·10	25·5	4·1	,, + 80% ,,	No attack	,,	,,	,,
7	0·10	25·7	6·1	,, + 60% ,,	,,	,,	,,	,,
8	0·10	25·4	8·45	,, + 40% ,,	,,	,,	,,	,,

TABLE IV.

FURTHER TESTS ON DUPLEX CHROMIUM-NICKEL STEELS.

No.	C, %	Cr, %	Ni, %	Structure after A.C. 1,050° C.	Intergranular Breakdown after Reheating	
					1 hr., 600° C.	50 hrs., 600° C.
9	0·09	20·2	9·2	Aust. + 5% Ferrite	Fairly severe	Severe
10	0·09	20·6	9·6	,, + 8% ,,	,,	Moderately severe
11	0·09	22·6	10·2	,, + 15% ,,	Moderately severe	None
12	0·10	22·3	8·4	,, + 20% ,,	,,	,,

On the basis of their observations, Scherer, Riedrich and Hoch summarised structural conditions existing in these higher chromium-nickel steels, after softening at 1,050° C., in the diagram reproduced in Fig. 16. Curve AB is part of that originally drawn by Strauss and Maurer[40] as representing the minimum nickel content to produce a stable austenite on cooling from high temperatures. Curve CD gives the minimum nickel content, as determined by Scherer, Riedrich and Hoch, required for a completely austenitic structure. Compositions to the right of this curve and above AB have duplex structures of austenite and ferrite, the amounts of the latter being indicated on the diagram.

It will be noted that in these tests on duplex steels—both with and without vanadium—Riedrich and his collaborators found that such steels—unless their content of ferrite was 40% or over—were liable to undergo intergranular attack after a short heating at 600° C., but obtained immunity on prolonged heating at the same temperature. This does not, however, represent the maximum benefit which can be obtained from duplex steels. If silicon be added to the extent of 1·5% or 2·0%,[41] and particularly if it is accompanied by a chromium content of 20–22%, steels are obtained with a ferrite content of 25% or so which are completely immune from intergranular attack. Thus steels containing—

C.	Si.	Cr.	Ni.
0·11	1·96	21·9	8·7
0·11	1·93	22·0	10·0

were free from intergranular corrosion after softening at 1,100° C., followed by heating for half an hour at 650° C. and then testing with the acid copper sulphate reagent; they were equally good if the softening temperature was raised to 1,250° C. Samples of the second cast were also heated for various periods between 30 mins. and 2 months at 600°–700° C.; all these samples were quite free from intergranular attack when tested in the acid copper sulphate reagent.

Additions of molybdenum which produce a similar degree of duplexity also give a reasonable freedom from intergranular susceptibility, thus the following steels:—

C.	Si.	Cr.	Ni.	Mo.
0·08	0·51	19·3	8·2	3·7
0·07	0·57	19·5	8·1	4·0

were quite sound after softening at 1,100° or 1,250° C., reheating for 30 mins. at 650° C., followed by immersion in the boiling acid copper sulphate solution for 72 hours.

Similar results as to the freedom from susceptibility of duplex steels containing silicon or molybdenum have been obtained by Payson,[42] and by Pfeil and Jones.[43]

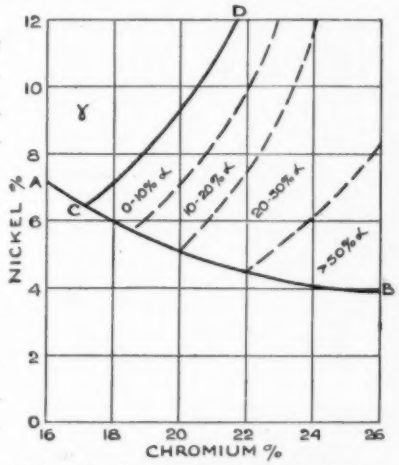

Fig. 16.—Structural conditions in chromium-nickel steels (Scherer, Riedrich and Hoch).
Curve AB—Minimum nickel content to give a stable austenite.
Curve CD—Minimum nickel content for completely austenitic structure.

Various explanations have been given for the beneficial action of ferrite in reducing or removing intergranular susceptibility. Payson[42] showed that when duplex steels containing silicon or molybdenum were reheated at 650° C. or thereabouts, carbide appeared to precipitate in the ferrite, in some cases forming a structure resembling pearlite. Photomicrographs of a silicon steel also indicated that this precipitation in the ferrite occurred preferentially at or near the ferrite-austenite boundary; the carbide was present, moreover, as separate particles and not as a continuous film. The author's observations on duplex steels containing silicon indicate that steels of this type, after being softened in the normal way and then heated for 30 mins.—e.g., at 650° C.—contain fine particles of carbide in the ferrite adjacent to the ferrite-austenite boundary. The carbide does not form a continuous film; it exists as individual particles just visible at a magnification of 600, after a light etching with an alcoholic solution of hydrochloric acid containing a small amount of ferric chloride, as shown in Fig. 17. It is not known whether these particles are chromium carbide or whether they contain some admixture of silicon carbide, but the steels containing them were quite free from intergranular susceptibility. If one assumes that the particles are in fact chromium carbide, freedom from susceptibility is explainable by the fact that the carbide precipitate is discontinuous and only forms at austenite-ferrite boundaries. Any chromium depletion which may have accompanied the formation of this carbide would be very localised and would certainly not form continuous boundary layers.

Scherer, Riedrich and Hoch considered that the beneficial effect of a duplex structure of ferrite and austenite in expediting recovery from intergranular susceptibility—it will be recalled that the steels they investigated were susceptible after heating for 1 hour at 600° C., but became

40 "Die Hochlegierten Chromnickelstähle als Nichtrostende Stähle," *Kruppsche Monatshefte*, Aug., 1920, p. 129.
41 British Patent 348,586, May 4/31.

42 "Prevention of Intergranular Corrosion in Corrosion-Resistant Chromium-Nickel Steels." P. Payson. *Trans. Am. Inst. Min. Met. Eng.*, vol. 100 (1932), p. 306.
43 "A Contribution to the Study of the Properties of Austenitic Steels." L. B. Pfeil and D. G. Jones. *J. Iron and Steel Inst.*, 1933, I, p. 337.

immune after heating at this temperature for 10 hours—was due to two causes :—

(*a*) The ferrite is richer in chromium than the austenite with which it is associated.

(*b*) Diffusion of chromium takes place much more rapidly in the ferrite than in the austenite.

The latter conclusion appears to have been drawn solely from the observed fact that susceptibility was removed from duplex steels of high chromium content much more quickly than from steels of similar chromium content to which sufficient nickel had been added to make them wholely austenitic. The possible effects of other factors on rate of recovery—e.g., different modes of carbide precipitation in the two types of steel—do not appear to have been considered.

Fig. 17.—C, 0.10%; Cr, 21.9%; Ni, 8.8%; Si, 2.4%; air cooled 1,050° C., reheated 650° C. for 30 minutes and air cooled; lightly etched showing carbide precipitated in the ferrite, x 600. This sample was quite free from inter-granular attack.

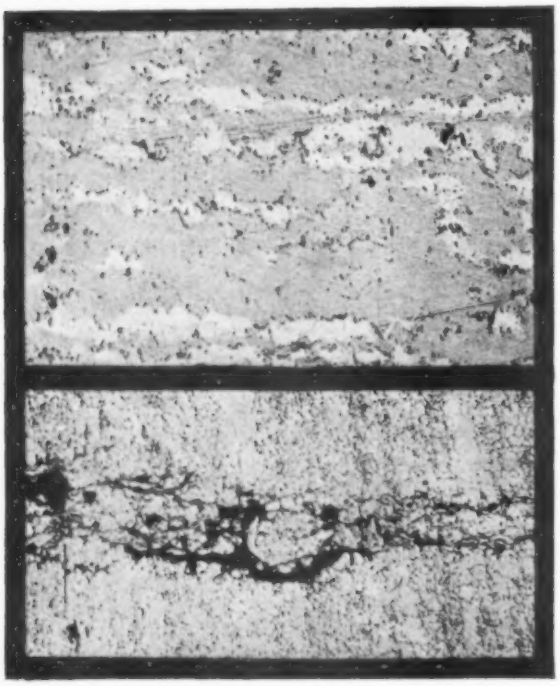

Fig. 18.—Intergranular attack on austenitic steel shrouding fitted to blading in a marine turbine. x 175.

Houdremont and Schafmeister[44] also refer to the beneficial effects of a duplex structure, and while not giving a detailed explanation, they state : " As the solubilities of ferrite and austenite for carbide differ, it was to be expected that, as a result of this solubility difference, the boundaries of the ferrite grains and the austenite grains would show differing amounts of precipitated carbide after annealing. By this means a beginning of intercrystalline corrosion could occur at individual grains, but the breakdown of the whole material through and through was prevented. At the same time the ferrite had apparently the peculiarity of causing a slight balling up of the carbide, mostly in the middle of the grain and less at the boundaries."

Steels which owe their immunity from intergranular attack to a duplex structure, such as is obtained by the addition of silicon or molybdenum, appear to have certain advantages as compared with steels which have obtained similar immunity by the addition of strong carbide-forming metals. Their resistance to intergranular susceptibility is generally not impaired, by softening at very high temperatures, to the extent which has been observed, for example, in titanium-treated steels. Again, silicon, though a deoxidiser widely used in steelmaking, is not oxidised as rapidly as titanium during welding operations ; weld metal reasonably immune from intergranular susceptibility can therefore be laid down from electrodes or filler rods of duplex silicon-containing steels of suitable composition. Such steels are also very suitable for the production of small castings where, as mentioned earlier, the progressive oxidation of titanium during casting operations may cause considerable trouble.

Duplex steels have, however, a number of disadvantages. The ferrite and austenite in equilibrium in a given steel cannot have the same composition ; the ferrite always contains less nickel and more chromium and other " loop-forming " elements which may be present. Whether this will have any serious effect on the general corrosion resistance of the steel will depend on many factors, but it may lead to either the ferrite or the austenite being preferentially attacked. It should be noted that the differences in composition between these two constituents in a given steel may be quite considerable.

A more important disadvantage arises from the fact that the alpha and gamma phases do not deform similarly in hot-working operations, consequently highly duplex steels have frequently poor hot-working qualities. They are liable to crack or fissure during forging or rolling, and frequently give a poor yield of sound material. This disadvantage does not arise, of course, when castings are in question.

In addition, some of the duplex steels are liable to embrittle to a greater or less extent if they are heated continuously at a dull red heat, due to the formation of the sigma phase. The temperature range in which such embrittlement may be produced varies with the actual composition of the steel, but may extend from about 400° to 900° C. or even somewhat higher. Where conditions of use involve prolonged heating under stress at such temperatures, it is, therefore, advisable to determine whether appreciable embrittlement is likely to occur with some particular steel whose use may appear desirable on account of freedom from intergranular susceptibility or on other grounds.

It may be noted that embrittlement may be produced very rapidly in some alloys, a few hours' heating at a dull-red heat being quite sufficient to reduce toughness and ductility to relatively low values ; in other cases, the change occurs only slowly, and even after prolonged heating the loss in ductility may not have serious consequences.*

It may be noted that austenitic chromium-manganese steels are susceptible to intergranular attack in a similar manner to austenitic chromium-nickel steels. Elimination of susceptibility in these steels does not appear to have been studied in such detail, but, broadly speaking, the same methods may be applied as with the nickel-containing steels. The problem is in certain respects more difficult on account of the fact that manganese is considerably less efficient than nickel as an austenite stabiliser in high-chromium steels. If chromium content exceeds about 15%, the chromium-manganese steels are not completely austenitic (unless carbon content is raised appreciably), but contain considerable amounts of ferrite. If their carbon content is low they may possess a sufficiently long " safe period " to allow straightforward welding without developing intergranular susceptibility in the neighbourhood of the weld. If, however, adequate amounts of carbide-forming metals, such as titanium or columbium, are added to obtain complete immunity against longer periods of heating at a dull red heat, the steels may become preponderatingly ferritic and hence lose much of the toughness and ductility associated with austenitic steels. Up to now, austenitic chromium-

44 " Prevention of Intergranular Corrosion in Steels containing 18% Cr. and 8% Ni." E. Houdremont and P. Schafmeister. *Archiv. fd. Eisenhüttenw.*, vol. vii, p. 187, Sept., 1933.

* A detailed account of the occurrence of the sigma phase was given in this journal, March, 1940, p. 143 and July, 1940, p. 87, to which those interested may be referred.

manganese stainless steels have not been used to any extent in this country, as they are in a number of respects inferior to austenitic chromium-nickel steels of like chromium content; one suspects that their use in Germany during recent years has not been due to the possession of particular virtues, but rather has been dictated by restrictions in the use of nickel.

Practical Considerations

The very many purposes, involving widely different conditions both of fabrication and of service, for which the austenitic chromium-nickel steels are used, make it likely that specific methods of combating intergranular susceptibility may not be equally successful in every instance; some indications of this have been given in the earlier parts of this article. On the other hand, methods may be completely satisfactory in eliminating any possibility of intergranular attack, but may be needlessly expensive for some particular purpose. Two vital questions, to be considered when deciding whether a given steel processed in a stated manner will be satisfactory for some particular purpose, are the corrosive conditions to which the finished article will be subjected and the maximum time the steel, after it has been finally heat-treated, will be heated in the sensitizing range of temperature, both during fabrication and in service.

The question of corrosive conditions is obviously important, because if corrosion is absent intergranular attack is not likely to develop in susceptible material. As an extreme case one may instance the use of "18-8" steels as non-magnetic material in electrical switch gear. Steel so used is frequently welded in position, but as the welded part is immersed in oil, no intergranular breakdown results, although the steel so used may be, and frequently is, a plain chromium-nickel steel containing up to 0·15% carbon and hence highly susceptible when heated at a dull red heat for even a short time. Another even more interesting example occurred in the low-carbon "18–8" tubes built into the experimental superheater at the Trenton Channel plant of the Detroit Edison Co., to which reference has already been made. These tubes were heated continuously during use in the range 540°–600° C.; they carried superheated steam under pressure and were in contact externally with the products of combustion of oil fuel. During a period of approximately two years they were in a highly susceptible condition, but no intergranular attack resulted because, presumably, the dry superheated steam was not sufficiently corrosive to attack the chromium-depleted layers. If one should have the idea of imitating this extremely interesting example, it would be well to make sure, however, that the corrosive conditions are exactly duplicated. Absence of attack at Detroit was almost certainly due to the steam being very pure and highly superheated. If it had been contaminated—e.g., with chlorides—and condensation had occurred, the result might have been different, as witness the example in Fig. 18, which shows intergranular attack in shrouding strip fitted to blading in a marine turbine. This strip was made of steel containing 0·14% carbon, 16·0% chromium, and 10·3% nickel, and had been cold-rolled to shape; during the final softening treatment it had not been cooled sufficiently quickly to prevent the separation of intergranular carbide. After being in use for some time, cracks developed in the shrouding and examination showed conclusively that they had formed as a result of intergranular attack. There seems little doubt that corrosion occurred as a result of chloride contamination of the steam.

If a welded article is subjected to very mildly corrosive conditions only, it is probable that intergranular corrosion will develop only very slowly in susceptible areas; it is conceivable it might not develop at all, though the fact that it has been found to result from contact with warm sea-water[45] or with boiling water containing little more than traces of sulphuric and nitric acids indicates that corrosive attack need not be intense to produce serious effects. At the same time, it is a fact that large quantities of ordinary "18–8" steel have been used for household and kitchen utensils, shop-fittings and decorative fixtures—both indoor and outdoor—which have been made from sheet and whose fabrication has involved welding. Some of the articles may have been heat-treated after welding, but it is certain that many hundreds have not, and one does not hear of failures from these due to intergranular attack. It may be that such welded articles are more liable to pit in the neighbourhood of the weld—this has been observed in certain polished fittings—but the danger of serious trouble does not seem to be great.

Frequently, spot-welding is used for such purposes, and in this case the probability of attack may be practically non-existent. The actual time required to make a spot weld in reasonably thin sheet may be only a fraction of a second, and as the material cools down very quickly as a result of heat conduction to the surrounding cold metal, there is not time for any appreciable precipitation of carbide. For example, spot welds on cold-rolled 20 G strip of steel containing 0·14% carbon, 16·2% chromium, and 10·5% nickel were quite free from intergranular attack when tested with the acid copper sulphate reagent. Thicker sheet would require current to be passed for a somewhat longer time, but even this would not be more than a second or two, so that, unless the steel being used was abnormally sensitive, no appreciable susceptibility should be developed. Similar remarks would also apply to other very rapid methods of welding—e.g., the butt-welding of strip.

In the brewing and dairy industries, corrosive conditions again are generally mild and, as mentioned in an earlier part of this article, steel with but a very limited resistance to intergranular breakdown has been used on the Continent for the construction of equipment for these industries. Much of such equipment, however, is of large scale and costly to construct; it would seem inadvisable, therefore, to risk the breakdown of such expensive equipment merely to save a very small fraction of its cost. The author would recommend that for such equipment one should select steels which have an ample "safe period" at 650° C. to cover any welding operation likely to be involved. Whether 30 mins. constitutes the minimum "safe period" which should be allowed, or whether a reduction to some shorter time—e.g., 5 or 10 mins.—could be permitted without undue danger depends on the thickness of the parts to be welded and the complexity or otherwise of the welded joints. There seems little doubt that 10 mins. would provide an ample margin of safety for most constructions involving sheet up to about 8 gauge.

A point which may have considerable importance in connection with attack by mild corrosives on steel which is somewhat susceptible is the presence or absence of internal stress in the steel. Intergranular attack takes place along very narrow paths. It seems probable, therefore, that attack will be helped if the steel is under tension and retarded if it is under compression. One would imagine that the penetration of the attacking liquor into the microscopically narrow channels it must traverse will be speeded up if there is tendency for these channels to open slightly; on the contrary, compressive stresses would tend to close a channel already formed and prevent further ingress of the attacking liquor.

An example, interesting in several ways, which seems to point in this direction came to the author's notice some years ago. A cylindrical storage tank for concentrated nitric acid erected in a chemical works in this country developed visible cracks on its external surface after a few months' use. The tank was of rivetted construction and the cracks ran parallel to the rivet holes in a vertical lap joint and about $1\frac{1}{8}$ in. from their centres. On the opposite side of the plate to the crack there had been heavy caulking on the edge of a second plate, riveted to the first, thereby causing a slight bulging of the outer surface of the latter; the cracks occurred at this slight bulge. The plates were $\frac{11}{32}$ in. thick and were made of "18–8" steel containing 0·19% carbon.

[45] "Stainless Iron and Steel," (2nd Edition), p. 281 and Fig. 169.

Microscopic examination of samples cut from the faulty plate both in the neighbourhood of the cracked joint and some distance away where the surface appeared to be sound, showed that the cracks visible to the eye were undoubtedly intergranular, that the whole external surface of the plate was full of microscopic intergranular cracks which opened out as soon as the plate was bent slightly and that the steel itself was in a highly susceptible condition, a very well-defined carbide network being present. The most interesting point, however, was that none of the cracks mentioned above penetrated beyond the middle of the plate; the inner surface of the plate and the inner half of its thickness were completely free from intergranular attack although the structure of this half was equally as bad as that of the outer half. This could be demonstrated by bending strips cut from the plate; those with the inner surface in tension bent satisfactorily, whereas those with the outer surface in tension cracked immediately. This may be seen from Fig. 19, which also shows that the cracks starting from the outer surface end sharply at the middle of the plate or before they reach that position.

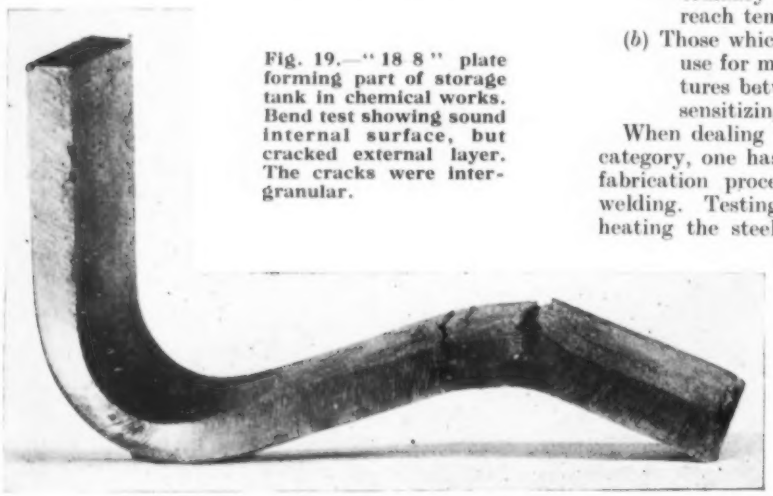

Fig. 19.—"18 8" plate forming part of storage tank in chemical works. Bend test showing sound internal surface, but cracked external layer. The cracks were intergranular.

During the fabrication of the vessel, the plate had been bent cold to the required cylindrical shape; the outer surface of the bent plate would therefore be left in tension and its inner surface in compression. So far as is known, the only other difference between the two surfaces was that externally the vessel had been exposed to the somewhat corrosive atmosphere of a chemical works, whereas internally, for part of the time, it had been in contact with cold concentrated nitric acid. While it may be agreed that the latter would not be expected to have any appreciable corrosive action on stainless steel, even with a chromium content of 10% or so, it appears extremely probable that the marked intergranular attack which proceeded from the external surface, and did not pass beyond the middle of the plate, had been greatly accelerated by the internal tension existing there. The complete absence of intergranular attack in the inner half of the plate thickness—where internal stress was in compression—is particularly noteworthy. The additional tension produced by the heavy caulking on the inner surface of the plate, which caused the outer surface to bulge slightly, would appear to account satisfactorily for the cracks first becoming visible in that bulged area.

When corrosive conditions are severe, every precaution should be taken to ensure that plant subjected to these conditions is free from areas susceptible to intergranular attack, otherwise failure may occur quite rapidly. This applies not only to the chemical industry, but also to others, including textile dyeing and certain branches of the food industry—e.g., those dealing with sauces, pickles and other products containing vinegar and common salt among their various ingredients. It has been the regular custom in this country for a number of years to test steels used for these purposes with the acid copper sulphate reagent, and as one does not hear of evidence to the contrary there seems little doubt that its indications are generally reliable. At the same time, there is a possibility, as was mentioned in the first part of this article, that certain very active chemicals, as, for example, boiling concentrated nitric acid, density 1·4, might cause accelerated attack on areas slightly depleted in chromium content which nevertheless successfully resisted the acid copper sulphate reagent. When plant has to withstand the action of such severely corrosive liquors, it would seem advisable to use the liquor itself as a test reagent for intergranular susceptibility, as well as the normal acid copper sulphate solution.

When considering suitable tests to determine whether given steels are likely to retain their immunity against intergranular attack, when used for specified purposes, it may be helpful to divide these purposes into two categories:—

(*a*) Those in which the completed vessel is used at ordinary temperatures, or, if heated in use, will not reach temperatures higher than about 350° C.

(*b*) Those which involve the equipment being heated in use for more or less prolonged periods at temperatures between about 400° and 800° C.—i.e., in the sensitizing range of temperature.

When dealing with equipment for purposes in the first category, one has to take into account only the effect of fabrication processes, of which the most important is welding. Testing from this point of view should involve heating the steel at 650° C. (after it has been properly softened) for a time which will adequately cover, with a margin of safety, any possible welding conditions which may arise in making the plant. In this connection, it should be remembered that the effects of repeated heatings in the sensitizing range are additive, and hence, in the welding of heavy vessels, one must take into account the number of runs required for a particular weld, the possibility that weld defects may have to be cut out and rewelded and that two or more welds may meet in one place. Whether 10, 30 or 60 mins. will adequately cover any such possible combination will depend on the size and complexity of a given plant, but it is always well to err on the safe side. Obviously, care should also be taken to ensure that the electrodes or welding rod used will deposit weld metal which will not be made susceptible if it is reheated to a dull red heat by a subsequent weld meeting or crossing it, a matter to which sufficient attention has not always been given.*

If tests on actual welds are preferred to those in which a sample of the steel is heated for a stated time to 650° C. or thereabouts, it is obvious that the welds to be tested should be made in a manner comparable to those of the actual equipment. The material tested should have the same thickness as that to be used in the plant, and there should be at least one pair of crossed welds. A straight single-run weld in thin sheet is of little value in assessing the suitability of a particular steel for equipment made from plate and involving multiple welds.

Other fabrication processes besides welding involve heating the steel, and the question frequently arises as to whether heat treatment is necessary or advisable after such processes. Each has to be considered on its merits, and perhaps the specific case of the dishing or flanging of plates may be taken as an example.

If a plate can be flanged or dished rapidly in one operation

* See Fig. 15 in the third part of this article. This journal, September, p. 141.

Continued on page 194.

Copper and Copper Alloys

By William Ashcroft

Copper, in an annealed condition, is easy to work, but its strength and hardness is increased by cold-working, to increase further its strength and hardness various elements are added, and these affect other properties. The number of elements used is very considerable, thus a very wide range of copper-base alloys is available, and the nomenclature applied to them is so varied that the task of the engineer or designer in making selections is not an easy one. Here the author briefly reviews the subject.

COPPER was one of the first metals to be used by man, indeed, archæologists tell us that there was a cultural phase conditioned by the use of copper, without an intentional alloy, for implements and tools, intermediate between the stone and bronze ages, though there appears to be no sharp line of demarcation between it and the bronze age. Since these early days many metals have been discovered and have been applied to the needs of mankind, but copper continues to play a vital part in all branches of engineering. Especially is this true with regard to electrical engineering, where its high electrical conductivity makes it a valuable material. In these modern times and under normal conditions it is probable that more than half the copper produced is used in the electrical industry, whilst the remainder is employed for a vast range of purposes and in every industry. In addition to the electrical industry, copper as a metal is used in shipbuilding, locomotive construction, domestic water pipes, and in the dairy, brewing, chemical, laundry and many other industries; but much of the copper produced is alloyed with varying percentages of other metals by which the properties of the copper can be adjusted to suit almost any purpose, and it is these copper-base alloys with which this article is more directly concerned.

In its annealed condition copper is soft and easy to work; its strength and hardness, however, can be increased by cold-working, thus, by hammering, rolling or drawing copper when cold its tensile strength can be more than doubled. To increase the strength and hardness further other elements are added to the copper to improve these properties. Thus, for instance, if it is desired to increase the strength without appreciably reducing the electrical conductivity of the copper, the addition of less than 1% of cadmium will be effective. Stronger alloys are obtained by the addition of other elements, the addition of 5% of tin, for instance, is sufficient to double the strength of copper, while the addition of a small percentage of beryllium makes it as strong as a high-grade steel, with an exceptionally high resistance to fatigue. In addition to the elements mentioned must be included zinc, lead, nickel, iron, phosphorus, arsenic, manganese, silicon, aluminium, and perhaps others that are added to copper for one reason or another. Since two or more of these elements may be added to copper in any one composition, it will be obvious that the number of copper alloys used runs into many hundreds and the nomenclature applied to them is so varied that the task of the engineer or designer in making selections is not an easy one.

Copper-base alloys embrace the ordinary copper-zinc alloys or brasses, the copper-tin alloys or bronzes, and many well-known special alloys comprising particular types of these two ranges, and alloys of unique composition carrying their own specific designation, such as the cupro-nickels and nickel silvers. Much work on copper alloys has been carried out by individual firms, and particular compositions have been developed and marketed by these firms under trade names which are sometimes associated in some way with the names of the firms responsible, represent an abbreviation of the composition, or indicate some outstanding feature of the alloy. The original bronzes are the copper-tin alloys, but the name bronze has become familiar in its application to other alloys that have somewhat similar properties and it has become common practice to distinguish these by using the name of the element added to the copper as the prefix to bronze, thus such bronzes are known as phosphor-bronze, manganese-bronze, silicon-bronze, aluminium-bronze, and beryllium-bronze.

Most of the alloying elements used have a considerable solid solubility in copper, and because of this fact they are particularly effective in producing useful modifications of the properties possessed by the base metal. With copper and nickel, for instance, the solubility is complete from one metal to the other, and this provides a group of copper-nickel alloys in which copper predominates. Next to nickel, the metal that dissolves most freely in solid copper is zinc—producing the solid solution known as alpha brass. When less than about 61% of copper is present in the alloy at room temperature the saturation point is reached and further addition of zinc introduces a second structural constituent into the alloy known as the beta solid solution. From about 61 to 54% of copper the alloys are intimate mixtures of these two constituents. Similar conditions are found in other copper-base alloys—namely, a tendency to form intermediate solid solutions, each of which represents a particular kind of structure. Generally, it is the first, or alpha, solution that is of greatest importance. One of the exceptions to this generalisation is the copper-tin alloys, in which tin in excess of 14% is sometimes used for castings of high strength, although 14% of tin is generally considered to be the limit of solid solubility in copper at room temperature.

When a brass or bronze composition is modified by the introduction of another element, the name of the brass or bronze is generally qualified either by the name of the element or by the effect produced. Thus, lead is added to brass or bronze to facilitate machining, hence the term free-cutting brass or bronze. In large amounts lead is added to produce the so-called bearing bronzes for use in heavy-duty bearings. Phosphorus may be added to deoxidise and free a fluid bronze from tin oxide, and although the content of phosphorus in the finished product may be practically nil, it is frequently known as phosphor-bronze, but the true phosphor-bronzes are alloys which contain sufficient phosphorus to have an appreciable influence on their properties. Manganese also may be added to purify the metal before casting, but the term manganese-bronze is more generally applied to a foundry brass of the mixed alpha-beta type, containing more manganese than is necessary to deoxidise the melt, together with a little tin, aluminium and iron.

Industrial Bronzes

Tin-bronze.—In common practice the name bronze is applied to designate copper alloyed with almost any element excepting zinc, which is the brass-forming element; however, the copper-tin alloys are the original bronzes. These latter may be divided into two classes, those that contain about 5% of tin, which are suitable for hot- and cold-working, and alloys containing 9% or more of tin which are almost invariably used for castings. The low-tin alloys, when rolled and drawn to spring temper, are very

strong and highly resistant to fatigue. Phosphorus is generally added as a deoxidiser, and very small percentages are usually present; this group is probably the nearest approach to straight copper-tin alloys used in practice.

The second class of copper-tin alloys, containing 9% or more of tin, develop the characteristic properties that determine their use as castings. Some of the alloys with a tin content of less than 20% may be hot-worked at a temperature of about 600° C. and cold-worked if quenched from a somewhat higher temperature; in the annealed and wrought condition, however, these alloys are not very different from a number of other copper-base alloys that can be worked more easily. There is quite a range of tin-bronzes used in the production of castings, and nearly all contain additions of other elements, such as zinc, phosphorus, nickel and lead, either with the object of improving casting properties, purifying the metal, or improving its physical properties. Probably the main casting bronzes are the gunmetals, which are based on the composition containing 88% copper, 10% tin, and 2% zinc. Variations of this composition are used in practice, in which lead is added, and, in some instances, small percentages of nickel; the tin content of these alloys is generally within the 8 to 10% range.

As already mentioned, the phosphor-bronzes are really tin-bronzes, containing from 10 to 14% of tin, in which the amount of phosphorus added is sufficient to clean the metal and leave in the finished product an amount of phosphorus ranging between 0·1 to 0·3%. Nickel may be added to these bronzes, and in some cases lead is a constituent, but zinc is regarded as an impurity and should not be present. Phosphor-bronze castings are especially suitable for machine parts which require a bearing surface; they are particularly useful for parts exposed to the action of seawater. Phosphor-bronze in wrought forms—i.e., as rod, sheet and wire—are used for many purposes in which toughness, resistance to corrosion, and non-magnetic properties are required. For such purposes the tin content rarely exceeds 8·5%, and may be as low as 3%.

Another series of tin-bronzes is used for heavy duty bearings and has a high lead content, varying between 8 and 20%. The tin content of these alloys is usually within 8 to 10%, and in addition to these two elements, zinc and nickel or nickel and phosphorus are almost invariably included in small amounts. Mention may also be made of the bell metals, which contain a relatively high precentage of tin, usually about 20%, but sometimes as high as 30%. Zinc and lead are invariably present in these tin-bronzes.

Aluminium Bronze.—The term "aluminium" is usually applied to an alloy containing from, say, 2 to 16% aluminium and the remainder copper, though small to appreciable percentages of other elements, such as iron, manganese, or nickel, are usually present. Diverse compositions, either in the simple binary or more complex alloys, combine remarkably good mechanical properties with excellent resistance to corrosion by many media, particularly to attack by sulphuric acid. Difficulties encountered in the production of castings restricted the application of aluminium bronze for many years, and apart from die-castings, it is not yet greatly favoured in the non-ferrous foundry. Dross, shrinkage, gas cavities, and isolated patches rich in aluminium oxide are troubles that are too common in the production of sand castings, although they are being produced in a regular commercial scale.

It is rather surprising that greater use is not made of aluminium bronze, because the commercial alloys of this type combine the strength and ductility of medium carbon steel with high resistance to corrosion. The alloys containing up to about 11% aluminium are readily hot-worked, but for cold-working the aluminium content should not exceed 7%, and may be much less. Most aluminium bronzes used for the production of sand or die-castings have been derived from modification of the 90-10 copper-aluminium alloy, but most aluminium bronzes now used are more complex. Other elements are added for various reasons. Thus, (1) to prevent self-annealing of the 9–11% aluminium alloys when slowly cooled; (2) to improve the mechanical properties; (3) to improve the resistance to certain corrosive media; (4) to improve the casting or working qualities; and (5) to increase the susceptibility to heat-treatment. Probably the most widely used of these elements is iron, while manganese, which is generally used as a deoxidiser, is usually present to about 0·5%, though in some cases this may be about 3%; nickel also is frequently added, while other elements present, such as tin, zinc and silicon, may be regarded as impurities, although in certain alloys silicon is beneficial.

Silicon-Bronze.—This is another alloy which combines high strength with resistance to corrosion, and which can be hot- and cold-worked. The silicon content usually is between 1·5 and 4·0%, and manganese is generally present, while small percentages of tin, zinc, and lead may be present. Silicon is the principal alloying addition to several proprietary copper-base alloys.

Brasses

Of the copper-base engineering alloys in demand to-day by far the greater weight is comprised of the copper-zinc base alloys or brasses. These alloys are divided into two main groups—the high-tensile and the lower tensile brasses. The lower strength alloys consist in the main of the binary alloys of copper and zinc, usually in the proportion of 60% copper, 40% zinc; or 70% copper, 30% zinc, with or without the addition of small percentages of tin and other elements. These alloys are commonly referred to as naval brass, cartridge brass, condenser brass, Muntz metal, delta metal, etc. The high-tensile brasses are more complex in composition and, in addition to the copper-zinc base, may contain aluminium, tin, iron, manganese and nickel in varying amounts. A typical example is the alloy frequently referred to as manganese bronze, but which is more correctly named manganese brass, since it is a complex alloy of the 60 copper 40 zinc type, to which comparatively small percentages of other metals are added, replacing in part either the copper or zinc, as the case may be. These additional metals include tin, manganese, aluminium, and nickel, which, in a properly compounded manganese brass, confer upon the alloy greatly enhanced mechanical properties, resulting in these complex brasses being in great demand for parts subjected to severe duty. Manganese brass may be forged, stamped, rolled, extruded or drawn, as well as used in the "as cast" condition, and its strength may be further improved by suitable heat-treatment.

The solid solution of zinc in copper, known as alpha brass, can contain about 38% of zinc at ordinary room temperature, and since alpha brass possesses a very high degree of plasticity, it is easily extruded, forged, rolled, pressed and drawn, further it is readily cast in intricate shapes and is easily machined, soldered or otherwise fabricated. Generally, the strength and ductility of the copper-zinc alloys increase with the zinc content within the alpha range, those considered to have the highest ductility are in the region of 70 copper 30 zinc; these are subjected to the most severe drawing operation, and a typical alloy of this type is known as cartridge metal.

The copper-zinc alloys which contain both alpha and beta solid solutions, which are represented by the typical alloy 60 copper 40 zinc, form a group of alloys containing about 61 to 54% copper. The outstanding feature that distinguishes these alloys from the alpha brasses is their extreme plasticity at red heat, followed by conditions tending towards brittleness when cooled to room temperature. They can be hot-worked, but are not suitable for severe cold-working. When heated to a suitable temperature, billets of a composition within this range can be extruded into rod or other shapes at a minimum expenditure of energy. Many operations are carried out by combining

not-working in the breaking-down stages, followed by a series of cold-working and annealing operations substantially as would be conducted with alpha brasses.

All the alpha-beta brasses are predominantly beta when brought to a bright red heat, because of the decreased solubility of zinc in copper which results in a transformation from alpha to beta. In this condition the alloys are very plastic, but when cooled down the 60 copper 40 zinc composition normally contains about 85% of the alpha phase which enables it to be cold-worked; if it is required to be worked a bit more freely the copper content may be increased. On the other hand, if the copper content is reduced to about 56% the resulting alloy works admirably at red heat, and is hard and strong when cooled, but it cannot be worked satisfactorily in the cold state.

It is not possible to discuss the diverse uses of the many brass compositions, but some idea of this diversity may be gathered from an appreciation of the fact that the many compositions can be produced with different degrees of work-hardening, varying from soft to spring temper, and can be further strengthened by adding tin, or improved from a machining point of view by adding lead. The properties of the material, therefore, depends not only upon the composition, but upon its condition at the time of testing, and in cast alloys the condition depends upon whether it is sand-, chill- or die-cast and upon the thickness of metal in the casting. In forged and annealed alloys the properties vary with the degree of work put into them, the temperature at which the work is carried out, and the time and temperature of annealing.

Effect of Added Elements on Brass

Mention has already been made of the fact that one or more metals may be added to copper-zinc alloys for a variety of reasons, and it is of interest to consider briefly the effect of these elements on the base alloy. Most of the elements used have considerable solubility in the base alloy, and produce useful modifications in properties, increasing the variations possible to meet particular requirements.

Aluminium is added to brass to improve its resistance to corrosion and to increase its strength. It tends to retard the oxidation of the zinc and thus produces cleaner castings, but care must be taken in pouring the metal. Advantages in the use of aluminium are shown by the resistance to corrosion of condenser tubes containing this element. The alloy for this purpose is generally about 76 copper, 22 zinc, and 2% aluminium, and it is interesting to note that in some recent tests on twelve different alloys used for condenser tubes the alloys containing aluminium had the lowest percentage loss of weight. A study of the aluminium brass samples indicated that the optimum aluminium content is near to 2·5%.

The addition of aluminium to, say, a 70–30 brass, may be made with a view to increasing both the strength and ductility, or an increase in strength at the expense of ductility. Up to about 1% aluminium increases both strength and ductility; increasing the aluminium to 2% further improves strength, but at the expense of ductility; by raising the copper content, however, to 78% and retaining the 2% of aluminium, the ductility is improved at the expense of reduced strength. Generally, it can be stated that aluminium is an important constituent in high-tensile brasses.

Like aluminium, the addition of tin to brass increases its strength and resistance to corrosion, but somewhat similar results may be obtained by the addition of silicon, and it is not now used to so great an extent as formerly, although in Admiralty brass the base 70–30 alloy contains 1% of tin at the expense of a similar amount of zinc, and in naval brass tin is included, the composition being 62 copper, 37 zinc, and 1% tin. Small percentages of tin are often included in the complex high-duty brasses, while some of the red brasses included as much as 2%. Several other engineering copper-zinc alloys include tin, some are regarded as low bronzes, such as the 85 copper and 5% each of zinc, tin and lead.

Apart from the alloys of copper and nickel, a typical example of which is Monel metal, nickel is more commonly used as a constituent in bronzes, in which its use is beneficial in many directions. By replacing part of the tin with nickel a more economical alloy is produced with improved mechanical properties, particularly increased ductility and toughness, but the greatest benefit is obtained by suitable heat-treatment. Those bronzes which contain up to about 5% of nickel are known as low nickel bronzes, but there is also a very useful range of high-nickel bronzes in which the nickel content varies from 20% to an amount in excess of the copper content. These alloys are used to meet special conditions in service and resist the effects of corrosive waters and steam.

Nickel is also a constituent in some of the brasses. In certain compositions its use results in an increase in strength without loss of ductility; it also improves resistance to corrosion of the brass and the retention of strength at high temperatures. The nickel silvers also have as their base the copper-zinc alloy to which is added nickel in varying amounts from 5 to 30%. In these alloys the zinc content usually varies with the nickel, a typical example is 64 copper, 18 zinc, and 18 nickel to which a small percentage of manganese may be added.

Manganese is an effective deoxidiser of brass, and when only traces remain its use improves the properties of the metal. Larger amounts, such as are present in manganese bronzes, further increase the strength of copper-zinc alloys, in which the zinc content is within 39 to 45%, but rarely does the manganese content of manganese brass exceed 3%, more often it is about 1%. Iron, too, is an important constituent in high-duty brass; within certain limits its effect is to increase the strength of the metal, but it is used to increase ductility rather than to improve strength.

In referring to the addition of manganese, some mention should be made of the copper alloys containing silicon and manganese, the silicon contents varying from 3 to 4% and the manganese up to 1%; up to 2% of zinc and iron may be present in these alloys and possess good mechanical properties in the forged or otherwise worked condition. They possess exceptional properties at low temperatures, their strength and ductility both increasing with fall in temperature.

It has only been possible to refer briefly to a few of the more commonly used copper alloys, but elsewhere in this issue are a number of tables, giving the composition and properties of a number of them, and these will be supplemented by further tables in succeeding issues. The tables will be especially useful to engineers and designers in their desire to satisfy specification requirements.

United States Copper Production Increased

United States mines produced copper at an increased rate in the first five months of 1941, and if the rate for May is continued throughout the remainder of the year, production in 1941 will show a 12% increase over that in 1940, according to the Bureau of Mines, United States Department of the Interior. Production in 1940 was 878,086 short tons. Production in January-March, 1941, totalled 232,707 tons, production in April 80,113 tons, and production in May 83,683 tons. Substantial increases in production are indicated for Arizona and Utah and smaller increases for Montana and Nevada; a slight increase for New Mexico is also indicated.

Classification of Copper and Copper Alloys

By H. J. Miller, M.Sc.

(*Metallurgist, Copper Development Association*).

Copper-base alloys embrace such a large number of different compositions and their commercial nomenclature is so varied that the whole industry might well standardise on fewer compositions. Under present conditions, however, the method of classifying copper and its alloys, given by the author, will greatly assist engineers and designers in their task of specifying materials.

THE enormous range and varieties of copper and copper-base alloys which are in commercial usage has in the past always presented a problem to engineer-users and designers faced with the task of specifying materials, and at the present time when many firms are engaged in the fabrication of products and materials with which they are not fully familiar, the difficulties of the situation are more widely experienced. Tabulated data concerning the more important copper-base materials, with a citation of appropriate British specifications, will therefore probably be helpful at the present time.

It has naturally been impossible to include particulars of every alloy in commercial use, especially with proprietary materials, and it has also not been possible to give more than the most elementary details of properties, which have therefore been limited to tensile strength and related properties, and diamond pyramid hardness values of materials for annealed and work-hardened conditions, where applicable.

In view of the existence of many strange misconceptions on the subject, it might be pointed out that with the vast majority of commercial copper-base alloys, the application of cold work is the sole method of hardening and strengthening; the only materials to which heat-treatment may be applied are alloys of the precipitation hardening class, characterised by a diminishing alpha solubility range as the temperature falls, and the duplex alloy group, as instanced by aluminium bronzes, which are characterised by an

TABLE I.

COPPER—RODS, PLATES, SHEET, STRIP, TUBES AND WIRE.

Description.	Composition (%).			Typical Mechanical Properties.					British Standards or other Specifications.	Remarks.
	Cu.	As.	Other Elements.	Limit of Prop., Tons/sq. in.	0·1% Proof Stress, Tons/sq. in.	Tensile Strength, Tons/sq. in.	Elongation % on $4\sqrt{A}$.	Diamond Pyramid Hardness.		
H.C. Copper (tough pitch)	99·9+	—	—	1 (A)–12 (H)	4 (A)–21 (H)	14 (A)–24 (H) wire up to 30 (H)	4 (H)–55 (A)	50 (A)–110 (H)	Notes 1 and 2	Used for electrical and other purposes where high conductivity is essential.
Deoxidised Copper (phosphorised)	99·85+	—	0·015–0·1 P.	1 (A)–12 (H)	4 (A)–21 (H)	14 (A)–24 (H)	4 (H)–60 (A)	50 (A)–110 (H)	—	Almost invariably used in tube production, and where welding is necessary.
Arsenical Copper (tough pitch)	99·2+	0·3–0·5	—	1 (A)–12 (H)	4 (A)–21 (H)	14·5 (A)–24 (H)	4 (H)–55 (A)	55 (A)–110 (H)	Note 3	Used for applications involving flame heating owing to superior resistance to scaling and corrosion.
Deoxidised Arsenical Copper	99·2+	0·3–0·5	0·02–0·1 P	1 (A)–12 (H)	4 (A)–21 (H)	14·5 (A)–24 (H)	4 (H)–55 (A)	55 (A)–110 (H)	Note 3	Employed where welding is necessary and in tube production.
Copper (of unspecified Composition) for general purposes.	99·2+	—	—	As above for arsenical copper					899 (sheet) 840 (conduit)	
Silver-bearing Copper	99·9+	—	0·03–0·07 Ag	As above for H.C. Copper					—	The silver content increases the annealing temperature range at the same time having a negligible effect on conductivity.
Oxygen free and deoxidised H.C. Copper	99·9+	—	—	As above for H.C. Copper					—	Prepared under special conditions to ensure maximum conductivity and absence of oxide. Has greater ductility than tough pitch copper.
Free-cutting Copper	99	—	Up to 1 tellurium or selenium	1 (A)–10 (H)	4 (A)–18 (H)	14 (A)–22 (H)	10 (H)–45 (A)	50 (A)–110 (H)	—	Can be machined at maximum machining speeds; with correct production procedure the electrical conductivity will be over 95%.

NOTES: (1) British Standards 444 and 518 for strip, bars and rods; 159 for busbars; 445 for commutator bars; 23 for trolley wire sections.

(2) British Standards 128 for annealed H.C. wire; 125 for hard-drawn wire conductors; 174 for hard-drawn telephone wire; 176 and 177 for jointing wire.

(3) British Standards 24, Part 5, for locomotive plates, rods and tubes; 61 for heavy gauge tubes; 659 for light-gauge tubes; 5T7 and 2T51 for aircraft tubes.

TABLE II.

COPPER CASTINGS AND SPECIAL COPPER ALLOY CASTINGS FOR CONDUCTIVITY PURPOSES.

Description.	Composition (%).		Typical Mechanical Properties.					British Standard or other Specifications.	Remarks.
	Cu.	Other Elements.	Limit of Prop., Tons/sq. in.	0·1% Proof Stress, Tons/sq. in.	Tensile Strength, Tons/sq. in.	Elongation % on $4\sqrt{A}$.	Diamond Pyramid Hardness.		
Copper Castings (*a*) for high-conductivity purposes (*b*) for general purposes (*c*)	99·9 99·0 98·5	Deoxidants only 0·9 Cd and deoxidants 1 Sn, 0·5 other elements	—	2	10	25	35–40	—	(*a*) and (*b*) can be made to possess over 85% electrical conductivity. (*c*) usually possesses less than 50% conductivity, and is used for tuyeres and similar applications.
Copper-Chromium Castings	99–99·5	0·4–0·8 Cr and deoxidants	9	15	22	20	110	—	Must be heat-treated to give the properties recorded; otherwise it may be as weak as above materials. Electrical conductivity exceeds 80%.
Copper-Cobalt-Beryllium castings	97	2·6 Co 0·4 Be	20	32	42	10	240	—	A heat-treatable alloy giving over 45% conductivity and having excellent properties at elevated temperatures.
Copper-Cobalt-Silicon castings	97	2·5 Co 0·5 Si	20	30	40	10	220	—	Better than preceding alloy for casting.
Copper-Nickel-Silicon castings	97·5	2 Ni 0·5 Si	20	27	34	10	220	—	A heat-treatable alloy, capable of giving over 40% conductivity.

increasing alpha solubility limit with falling temperature.

With all materials the manufacturing sequence, and especially the last annealing and cold-working treatments, has an important influence on various other properties—for example, grain size, directional features, surface quality and suitability for deep drawing and pressing, all of which are beyond the scope of this review.

The various tables which accompany the present section of the article are devoted to copper in various wrought forms (Table I), copper castings (Table II), copper-rich conductivity alloys (Table III), and the commercial brasses and manganese bronzes in the forms of cold-rolled strip and sheet (Table IV), hot-rolled plate (Table V), wire (Table VI), tubes (Table VII), rods and sections (Table VIII), and castings (Table IX). It is intended in a later issue to deal with other copper-base alloys.

The data presented calls for little comment, but it should be realised that modifications of compositions and properties are favoured by different producers, this especially being the case with the more complex brasses, where the number of alloying elements is considerable. The development of the special brasses has been associated with the initiative of individual firms, who naturally have their own preferences, and this renders standardisation a difficult matter.

In the columns relating to mechanical properties, the

TABLE III.

COPPER ALLOYS FOR CONDUCTIVITY APPLICATIONS.

Description.	Composition. (%)		Typical Mechanical Properties.					British Standard or other Specifications.	Remarks.
	Cu.	Other Elements.	Limit of Prop., Tons/sq. in.	0·1% Proof Stress, Tons/sq. in.	Tensile Strength, Tons/sq. in.	Elongation % on $4\sqrt{A}$.	Diamond Pyramid Hardness.		
Cadmium Copper (*a*) 0·4% Cd (*b*) 0·9% Cd	99·5 99	0·4 Cd 0·9 Cd	2 (A)–15 (H)	4 (A)–30 (H)	18 (A)–35 (H) wire up to 45 (H)	5 (H)–60 (A)	60 (A)–130 (H)	23, 175, 672	Composition is left to discretion of manufacturers, the only requirements of the specifications being in respect of electrical conductivity, which varies in different specifications from about 82% for heavily worked material to 92% for lightly worked material.
Conductivity Bronzes (*a*) (*b*) (*c*)	99·5 99 99	0·5 Sn 0·5 Cd. 0·5 Sn 1 Sn	2 (A)–15 (H)	4 (A)–30 (H)	18 (A)–38 (H) wire up to 45 (H)	5 (H)–60 (A)	60 (A)–130 (H)	23, 175	B.S. 23 for trolley wire does not specify the composition, but calls for a minimum conductivity of 45%.
Copper-Chromium	99–99·5	0·5–1 Cr	15	25	35	15	140	—	Quoted properties are for heat-treated rods and higher values may be obtained in other conditions. Electrical conductivity exceeds 80% and the material may be used at temperatures up to 350° C. without impairment of properties.
Copper-Cobalt-Beryllium	97	2·6 Co 0·4 Be	20	42	48	15	240	—	Quoted properties are for heat-treated rods. The electrical conductivity exceeds 50% and service temperatures in excess of 400° C. may be used.

TABLE IV.

BRASSES—COLD-ROLLED STRIP AND SHEET.

Description.	Composition. (%)					Typical Mechanical Properties.					British Standards or other Specifications.	Remarks.
	Cu.	Zn.	Pb.	Sn.	Other Elements.	Limit of Prop., Tons/sq. in.	0·1% Proof Stress, Tons/sq. in.	Tensile Strength, Tons/sq. in.	Elongation % on $4\sqrt{A}$.	Diamond Pyramid Hardness.		
Cap Coppers ..	97	3	—	—	—	1 (A)–12 (H)	4 (A)–22 (H)	15 (A)–28 (H)	4 (H)–50 (A)	60 (A)–130 (H)	—	Industrial use practically confined to caps for ammunition.
	95	5	—	—	—	1 (A)–12 (H)	4 (A)–25 (H)	16 (A)–30 (H)	4 (H)–50 (A)	60 (A)–140 (H)	—	
Gilding Metals	90	10	—	—	—	2 (A)–15 (H)	5 (A)–30 (H)	18 (A)–33 (H)	4 (H)–55 (A)	60 (A)–150 (H)	713	Used for architectural metalwork, imitation jewellery, etc., on account of golden colours; various applications necessitating better corrosion resistance than is provided by brasses of lower copper content.
	85	15	—	—	—	2·5 (A)–16 (H)	4·5 (A)–30 (H)	19 (A)–36 (H)	4 (H)–60 (A)	65 (A)–160 (H)	712	
	80	20	—	—	—	2·5 (A)–17 (H)	4·5 (A)–32 (H)	20 (A)–40 (H)	5 (H)–65 (A)	65 (A)–170 (H)	711	
Brasses—												
Cartridge ..	70	30	—	—	—	3·5 (A)–18 (H)	5 (A)–33 (H)	21 (A)–45 (H)	5 (H)–70 (A)	65 (A)–185 (H)	267	Deep drawing brass having maximum ductiltiy in the range of brass.
65/35 ..	65	35	—	—	—	4 (A)–18 (H)	6 (A)–33 (H)	21 (A)–45 (H)	4 (H)–65 (A)	65 (A)–185 (H)	266	A good cold-working alloy.
Basis ..	63	37	—	—	—	4 (A)–18 (H)	6 (A)–35 (H)	22 (A)–47 (H)	4 (H)–55 (A)	65 (A)–185 (H)	265	General purpose alloy suitable for forming operations.
Yellow or Muntz metal (cold-rolled finish)	60	40	—	—	—	5 (A)–12 (H)	8 (A)–25 (H)	25 (A)–35 (H)	5 (H)–45 (A)	75 (A)–150 (H)	264	Will withstand only limited amounts of cold work.
Engraving Brass, typical composition.	62·5	36	1·5	—	—	4 (A)–10 (H)	6 (A)–20 (H)	21 (A)–35 (H)	5 (H)–55 (A)	65 (A)–150 (H)	—	Suitable for machining (shearing and notching) and engraving; limited capacity for bending.
Special alloy Brasses (*a*)	83	14	—	—	1 Ni, 1 Si, 1 Al	5 (A)–18 (H)	8 (A)–35 (H)	25 (A)–45 (H)	4 (H)–55 (A)	75 (A)–185 (H)	DTD 283A	Annealed materials used for aircraft purposes.
(*b*)	85–95	4–13	—	—	1–2 Si	4 (A)–15 (H)	5 (A)–30 (H)	21 (A)–40 (H)	5 (H)–60 (A)	65 (A)–175 (H)	DTD 263 and 267	Used for aircraft purposes. Will weld easily by resistance and other methods.

TABLE V.

BRASSES—HOT-ROLLED PLATE.

Description.	Composition. (%)					Typical Mechanical Properties.					British Standards or other Specifications.	Remarks.
	Cu.	Zn.	Pb.	Sn.	Other Elements.	Limit of Prop., Tons/sq. in.	0·1% Proof Stress, Tons/sq. in.	Tensile Strength, Tons/sq. in.	Elongation % on $4\sqrt{A}$.	Diamond Pyramid Hardness.		
Yellow or Muntz metal.	60	40	—	—	—	4	7	24	40	75	—	—
Naval Brass ..	62	37	—	1	—	5	8	25	40	80	409	Fairly resistant to corrosion.
Manganese Bronze—typical composition	59	38	—	1	1 Fe 0·2 Al 0·5 Mn	6	10	28	30	100	—	—

TABLE VI.

BRASSES—WIRE.

Description.	Composition. (%)					Typical Mechanical Properties.					British Standards or other Specifications.	Remarks.
	Cu.	Zn.	Pb.	Sn.	Other Elements.	Limit of Prop., Tons/sq. in.	0·1% Proof Stress, Tons/sq. in.	Tensile Strength, Tons/sq. in.	Elongation % on $4\sqrt{A}$.	Diamond Pyramid Hardness.		
Commercial Brass Wire, including pin wire, screw wire, rivet wire. (For nipple wire, see Table VIII).	61–70	30–38	0–0·5	—	—	—	—	21 (A)–50 (H)	1–60	65–200	—	Alloys of the higher copper contents are more ductile and suitable for severe cold-forming, such as heading. The lead content has a critical influence in certain applications.
Special Alloy Brass	83	14	—	—	1 Ni 1 Si 1 Al	—	—	30 (A)	—	—	DTD 367	Used for rivets for aircraft purposes.

suffix (A) refers to the annealed conditions, while the suffix (H) refers to the most fully work-hardened conditions which can usually be achieved in commercial production. The application of the heavy amounts of work required to achieve these latter properties are dependent to a large extent on the sectional dimensions.

TABLE VII.

BRASSES—TUBES.

Description.	Composition (%).					Typical Mechanical Properties.					British Standards or other Specifications.	Remarks.
	Cu.	Zn.	Pb.	Sn.	Other Elements.	Limit of Prop., Tons/sq. in.	0·1% Proof Stress, Tons/sq. in.	Tensile Strength, Tons/sq. in.	Elongation % on $4\sqrt{A}$.	Diamond Pyramid Hardness.		
85/15 Brass (Red Brass)	85	15	—	—	—	2·5 (A)–12 (H)	4·5 (A)–24 (H)	19 (A)–32 (H)	10 (H)–60 (A)	60 (A)–150 (H)	—	Used for condenser and cooling units, gauges and instrument tubes; not susceptible to season cracking. The composition is frequently modified by tin additions.
70/30 Brass....	70	30	—	—	—	3·5 (A)–15 (H)	5 (A)–28 (H)	21 (A)–38 (H)	10 (H)–70 (A)	65 (A)–175 (H)	378, 885, 886	Standard compositions for condenser tubes. The Admiralty brass has better corrosion resistance.
70/29/1 Brass (Admiralty Brass)	70	29	—	1	—	3·5 (A)–15 (H)	5 (A)–28 (H)	21 (A)–38 (H)	10 (H)–70 (A)	65 (A) 175 (H)		
Aluminium Brass.	76	22	—	—	2 Al	4 (A)–16 (H)	7 (A)–30 (H)	24 (A)–40 (H)	8 (H)–70 (A)	65 (A)–175 (H)	—	Possesses excellent corrosion resistance, and is a favoured alloy for condenser tubes.
Muntz metal ..	60	40	—	—	—	4	7	24	40	75	—	Mostly used for unimportant services.
Naval Brass ..	62	37	—	1	—	5	8	25	40	80	—	Has better corrosion resistance than Muntz metal.
Manganese Bronzes—(*a*) typical compositions	58	39	—	1·4	1 Fe 0·4 Al 0·2 Mn 0·7 Mn 0·8 Fe	8	15	33	25	120	—	These alloys combine high strength with fair corrosion resistance. Worked by hot methods and obtainable in heavy gauges only.
(*b*)	60	37·5	—	1		7	14	30	25	100		
Free-cutting Tube Brass	63	35·5	1·5	—	—	3·5 (A)–10 (H)	5 (A)–20 (H)	22 (A)–35 (H)	10 (H) 50 (A)–	65 (A)–150 (H)	—	The lead content is responsible for diminished ductility.
Special Alloy Brasses (*a*)	83	14	—	—	1 Ni 1 Al 1 Si	—	—	30 (A)	—	—	DTD 253A and 323	Used for aircraft service pipes.
(*b*)	83	13	—	4	—	—	—	25 (A)–40 (H)	8 (H)–60 (A)	70 (A)–175 (H)	DTD 318A	Used for aircraft purposes and condenser tubes.
(*c*)	85–95	4–13	—	—	1–2 Si	—	—	25 (A)–40 (H)	8 (H)–70 (A)	70 (A)–175 (H)	DTD 307 and 312	Will weld easily by resistance and other methods.

TABLE VIII

BRASSES—RODS AND SECTIONS.

Description.	Composition (%).					Typical Mechanical Properties.					British Standards or other Specifications.	Remarks.
	Cu.	Zn.	Pb.	Sn.	Other Elements.	Limit of Prop., Tons/sq. in.	0·1% Proof Stress, Tons/sq. in.	Tensile Strength, Tons/sq. in.	Elongation % on $4\sqrt{A}$.	Diamond Pyramid Hardness.		
Yellow or Muntz metal	60	40	—	—	—	4–7	6–12	22–30	15–40	75–140	—	The earliest hot-wrought brass from which other types have been developed. Will withstand a fair amount of cold working and bending.
Free-cutting Brasses (*a*)	58	39	3	—	—	4–9	7–12	25–33 (25 min.)	20–40	80–130	249	Most suitable material for high-speed machining, but can only be slightly deformed by bending, etc.
(*b*)	61	36	3	—	—	4–9	7–12	24–30	20–40	80–130	—	Known as American type of free-machining brass, and has greater impact strength than the British type.
Good Machining Brasses of superior impact strength. (*b*) (e.g., Nipple Wire Rod)	59	39·5	1·5	—	—	4–10	6–16	22–35	15–60	80–140	—	(*a*) is a standard hot-forging brass, which has also extended fields of application in extruded and cold-drawn conditions. (*b*) is a more ductile type of higher copper content.
	63·5	35	1·5	—	—	4–10	6–16	21–35	15–60	75–140	—	

TABLE VIII.—*Continued.*

BRASSES—RODS AND SECTIONS.

Description.	Composition (%).					Typical Mechanical Properties.					British Standards or other Specifications	Remarks.
	Cu.	Zn.	Pb.	Sn.	Other Elements.	Limit of Prop., Tons/sq. in.	0·1% Proof Stress, Tons/sq. in.	Tensile Strength, Tons/sq. in.	Elongation % on $4\sqrt{A}$.	Diamond Pyramid Hardness.		
Hot Forging Brasses (*a*)	58	40·5	1·5	—	—	4–6	6–8	22–28 (20 min.)	25–50	80–110	218	These possess good machining and stamping properties.
(*b*)	60·5	38	1·5	—	—	4–6	6–8	22–28	25–60	75–100	—	
Special Hot-forging Brasses (high tensile strength) —typical compositions (*a*)	58·5	39	—	1	1·5 Fe	7–10	12–16	27–33	25–50	90–130	—	Where machining operations are to be performed the presence of lead is desirable, and where soldering is involved the aluminium content should be less than 0·2%. If ultimate application is of a corrosive nature, alloys of about 1% tin content are mostly employed, and it is certainly advisable to avoid the all-beta brasses. The alloy (*f*) is a standard material in U.S.A., but little used in Great Britain.
(*b*)	59	38	0·5	1	1·5 Fe 1 Mn	7–10	12–16	27–33	25–50	90–130	—	
(*c*)	58·5	38	1	1	0·5 Fe 1 Mn	7–10	12–18	30–35	25–35	100–140	—	
(*d*)	57·5	38	1	1	0·5 Al 1 Fe 1 Mn	8–10	16–20	33–38	25–35	110–150	—	
(*e*)	57·5	37	—	—	1 Fe 1·5 Mn 3 Al	10–14	20–25	38–45	15–30	130–170	—	
(*f*)	65·5	—	—	—	3 Fe 4 Mn 4·5 Al	12–16	20–25	38–50	15–30	130–170	—	
Naval Brass— Ordinary ..	62	37	—	1	—	5–10	8–12	24–30	20–50	80–120	251	Used for structural applications and for forgings. The tin content is helpful in reducing corrosion susceptibilities especially in seawater. The leaded type of Naval Brass has a special application in stampings and forgings where machinability is essential.
Special ..	59	40	—	1	—	5–10	8–12	24–30	20–40	80–120	252	
Leaded ..	61	Rem.	0·5–2·0	1	—	5–7	7–10	24–30	20–40	80–120	—	
Brasses for architectural sections and similar purposes typical compositions (*a*)	58	40	2	—	0·4 Al	5–8	6–10	25–30	20–40	80–110	—	Hot-worked (extruded) surfaces have a bright yellow oxide film, due to aluminium content.
(*b*)	58	38	1	0·5	2 Mn 0·5 Fe	7–10	10–15	30–35	25–35	100–140	—	Hot-worked (extruded) surfaces have a chocolate brown colour, due to manganese oxide films.
(*c*)	44	44	2	—	10 Ni	7–10	12–18	32–38	20–35	100–150	—	A white alloy suitable for sections and also rods for stamping.
High Tensile Strength Brasses	Compositions as given above under special hot-forging brasses, together with modifications of these.					—	15 min.	30 min.	20 min.	—	250A	Some of the alloys are cold-worked to give the necessary strength, and hence if heated subsequently they show a loss of strength.
						—	18 min.	35 min.	15 min.	—	250B	
Ductile Brasses produced by cold working. (*a*)	70	30	—	—	—	3·5 (A)–12 (H)	5 (A)–25 (H)	21 (A)–35 (H)	10 (H)–70 (A)	65 (A)–160 (H)	—	Suitable for brazing: may be cold-headed.
(*b*)	80	20	—	—	—	3 (A)–12 (H)	4·5 (A)–25 (H)	20 (A)–35 (H)	10 (H)–70 (A)	65 (A)–160 (H)	3B11	

TABLE IX.

BRASSES—CASTINGS.

NOTE.—For common quality brass castings there are no generally recognised compositions in use in Great Britain, and the author has utilised below a classification adopted in U.S.A. for sand-cast alloys; other materials and British specifications are afterwards listed. The quoted elongation values in all cases are typical for separately cast test-pieces and not for cast-on test-pieces.

Description.	Composition (%).					Typical Mechanical Properties.					British Standards or other Specifications.	Remarks.
	Cu.	Zn.	Pb.	Sn.	Other Elements.	Limit of Prop., Tons/sq. in.	0·1% Proof Stress, Tons/sq. in.	Tensile Strength, Tons/sq. in.	Elongation % on $4\sqrt{A}$.	Diamond Pyramid Hardness.		
AMERICAN CLASSIFICATION OF MATERIALS.												
Leaded Red Brass (*a*)	85	5	5	5	—	—	5	14–20	15–35	60–70	—	Alloy (*a*) classified in Great Britain as a leaded gunmetal
(*b*)	83	7	6	4	—	—	5	13–17	15–27	55–65	—	
Leaded Semi-red Brass (*a*)	81	9	7	3	—	—	5	13–18	15–30	55–65	—	Good casting alloys capable of reproducing fine detail and largely used for ornamental purposes, also for low-pressure valves and fittings and similar products. British alloys usually have lower lead contents and do not adhere to close compositional limits.
(*b*)	76	15	6	3	—	—	4	13–18	20–35	55–65	—	

TABLE IX.—*Continued.*

BRASSES—CASTINGS.

NOTE.—For common quality brass castings there are no generally recognised compositions in use in Great Britain, and the author has utilised below a classification adopted in U.S.A. for sand-cast alloys; other materials and British specifications are afterwards listed. The quoted elongation values in all cases are typical for separately cast test-pieces and not for cast-on test-pieces.

Description.	Composition (%).					Typical Mechanical Properties.					British Standards or other Specifications.	Remarks.
	Cu.	Zn.	Pb.	Sn.	Other Elements.	Limit of Prop., Tons/sq. in.	0·1% Proof Stress, Tons/sq. in.	Tensile Strength, Tons/sq. in.	Elongation % on $4\sqrt{A}$.	Diamond Pyramid Hardness.		
Leaded Yellow Brass (*a*)	71	25	3	1	—	—	4	13–18	25–40	45–60	—	Alloys similar to (*b*) are well known in Great Britain and are used for production of ordinary commercial brass castings; their corrosion resistance is better than with alloys of lower copper content as (*c*).
(*b*)	66	30	3	1	—	—	4	12–17	20–35	45–65	—	
(*c*)	60	38	1	1	Al 0·1–0·2	—	5	16–20	15–25	55–80	—	
Leaded High-Strength Yellow Brass (manganese bronze)	61	35	0·75	0·75	— 1 Fe 0·75 Al 0·25 Mn	—	10–14	27–35	20–30	85–100	—	A typical all-purpose manganese bronze which can be machined fairly easily due to the lead content.
High-Strength Yellow Brass (manganese bronze) (*a*)	58	39·25	—	—	1·25 Fe 1·25 Al 3 Fe	—	12–16	32–39	25–35	95–130	—	In Great Britain many other compositions of manganese bronzes are in use, notably those strengthened by tin additions. (See below.)
(*b*)	62	26·5	—	—	0·25 Mn 5 Al 3·5 Mn	—	28–35	44–54	8–18	170–230	—	
OTHER MATERIALS.												
Brass Castings for Brazing.	85	15	—	—	—	—	4	12	40	50–60	—	—
Brass Die Castings— Ordinary Brass (*a*)	60	40	—	—	—	—	6	18–23	20–50	60–70	932	Small quantities of aluminium and silicon are usually present in these alloys.
Naval Brass (*b*)	62	37	—	1	—	—	7	20–25	20–50	60–80	920	
Common Brass Castings.	62–68	Rem.	2	1	—	—	4	12–17	20–35	45–65	—	Composition is very approximate.
Manganese Bronzes. (*a*)	58	Rem.	—	1	0·25 Al 0·25 Mn 1 Fe	—	14–16	32–36	20–35	95–130	—	See also compositions quoted above for leaded high-strength yellow brass and high-strength yellow brass.
(*b*)	58	Rem.	—	0·5	0·5 Al 1 Fe and Mn	—	14–16	32–36	20–35	95–130	—	
(*c*) 1	54 min.	—	0·5 max.	—	5 max.	—	12·0 min.	28 min.	25 min.	—		
2	54 min.	Rem.	0·5 max.	—	5 „	—	15·0 „	32 „	18 „	—		
3	54 min.	Rem.	0·5 max.	—	8 „	—	17·5 „	34 „	15 „	—	208	Generic Specifications.
4	50 min.	Rem.	0·5 max.	—	10 „	—	20·0 „	38 „	15 „	—		
5	50 min.	Rem.	0·5 max.	—	13 „	—	25·0 „	45 „	12 „	—		

Fatigue Tests on Some Copper Alloys

THE large amount of published data on the endurance properties of steel is in marked contrast with the paucity of information on copper alloys, and the determinations reported by Anderson and Smith* are therefore of considerable value. These determinations were made on carefully polished samples machined from commercially hard-drawn rods, and the values obtained are usually higher than those of investigators who utilised strip as commercially furnished in the rolled condition. A good range of well-known alloys has been tested, and the data give approximate ideal endurance strengths (at 300 million cycles in air) from which the effect of various factors, such as notches, corrosion and the like may later be evaluated. Pertinent information regarding composition, fabrication, structure and other properties is recorded.

The results seem to be well established for the particular alloys tested. The well-known effect of cold work in increasing endurance strength is shown for phosphor bronze and a silicon bronze. Comparatively little initial working raises the endurance strength rapidly, while further working produces less effect and may even reduce it, perhaps as a result of increased notch sensitivity. It is significant that "scatter" frequently increases with the amount of cold work.

Alloys whose structures vary with treatment show wide differences in endurance behaviour. As an example, naval brass is cited; it has an alpha-beta structure, and the amount and distribution of the two phases, greatly effected by treatment, itself affects the endurance properties. For the precipitation hardening alloys, the effect of precipitation heat-treatment is not nearly so great on endurance strength as on tensile strength, and considerable scatter of points may be obtained. This is indicated by curves for a heat-treated chromium copper alloy; although heat-treatment does not appreciably raise the endurance strength at 300 million cycles, it does greatly increase the life at high stresses. The results are extremely sensitive to slight variations in heat-treatment, which possibly accounts for much of the scatter. The S-N data for baryllium-copper also demonstrates the importance of heat-treatment variations.

* A. R. Anderson and C. S. Smith. Amer. Soc. Test. Materials. Annual Meeting at Chicago, 1941. (Preprint.)

Gravity Die-casting in Copper-base Alloys

By Arthur Street, Ph.D.

Slow progress has been made in the production of gravity die-castings in copper-base alloys. In this country there was a tendency to concentrate on the use of aluminium bronze for this purpose, because it proved to be a good die-casting alloy: progress, however, has now been made in the die-casting of brass, and some attempts to apply new alloys, but the whole field of die-casting high-duty copper-base alloys remains invitingly open for a good deal of extra work and progress in this country, as the author reveals in this survey.

IT is difficult to decide whether die-casting comes under the sphere of metallurgy or engineering. Probably the majority of those engaged in the industry look at it from an engineer's point of view. Thus, die-making and design and the various mechanical operations connected with the production, trimming and finishing of die-castings are well understood, but what might be called the "Metallurgy of Diecasting" is not given quite the same careful attention. Indeed, one hardly knows whether to be satisfied with the advances already made in the art of die-casting or impatient that still more progress has not been achieved.

Die-casters have not concerned themselves much with theoretical considerations on the method of flow of the liquid or pasty alloy when it is poured or injected into a metal die, and there has not been much work on the structure of the alloy in its die-cast condition. There has been no completely satisfactory explanation why some alloys are easy to die-cast while others are found to be more difficult. Then it would seem unfortunate that while general research and development work is progressing with, say, the copper-rich copper-aluminium alloys, containing additions of iron, nickel, chromium, zinc, beryllium, vanadium, etc., very little is heard of the die-casting of these new and promising materials.

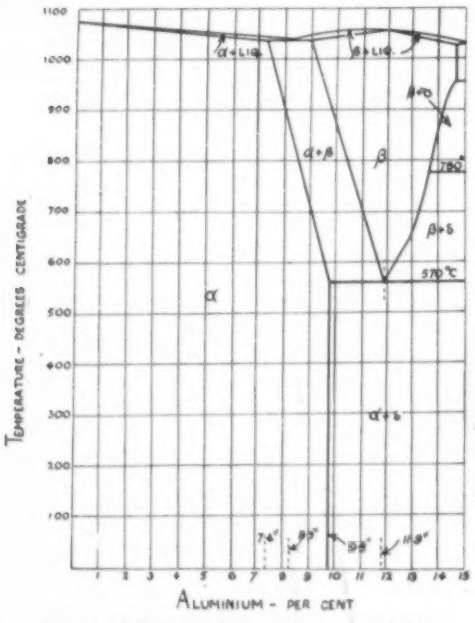

Fig. 1.—Equilibrium diagram of aluminium-bronzes containing up to 15% aluminium.

One trouble is that it is difficult to conduct an empirical research on die-casting alloys without getting so much back to first principles that the investigation appears to lose value when compared with modern production methods. For instance, the chill-casting of a test-bar does not teach us sufficient about the "die-castability" of complicated shapes in the alloy which is being tested. If a new alloy is developed which proves difficult to die-cast by existing methods, there is a tendency to ignore the potentialities of the new material and to continue producing the die-castings in the old well-established alloys—a stultifying influence on research which probably holds in many other fields besides die-casting.

Until about 1938 this was particularly noticeable in the die-casting of copper-base alloys where there had not been a great deal of development for many years. In this connection it is interesting to refer to a paper on the gravity die-casting of aluminium bronze by Rix and Whitaker in the *Journal of the Institute of Metals* for 1918.[1] In this paper, illustrations of gravity cast brush-holders are shown, quite as complicated as the majority of parts we produce to-day. The die material is given as close-grained, chill-cast iron, with cores of chrome-tungsten steel. "Die lives" for aluminium bronze are given as 5,000 to 7,000 castings, which compares quite favourably with the 10,000 averaged nowadays. For 20 years after the publication of that paper the only developments which occurred in copper alloy gravity die-casting were the production of bigger parts, slight improvement in die life, a fair amount of progress in the gravity casting of brass, and some (rather half-hearted) attempts to die-cast new alloys.

One reason for this apparent slow progress is that in aluminium bronze the early die-casters dropped on to a very good alloy and so it is not surprising that they concentrated their effort on finding new applications for die-castings in this material rather than in experimenting with other copper-base alloys or modifications of aluminium bronze itself.

Aluminium Bronze

The alloy of this type which is usually die-cast is composed of copper with 9–10% aluminium and up to 2% of iron. Even 25 years ago it was realised that the iron addition was advantageous, as it refined the grain of the alloy and considerably reduced the tendency of the aluminium bronze to embrittlement after slow cooling (although it must be said that in any case a die-casting is so rapidly chilled in production that troubles brought about by slow cooling can largely be ignored). As regards structure, the normal iron additions have little effect, so the following brief description of the constitution of straight aluminium bronze may equally be applied to the alloy containing up to 2% iron.

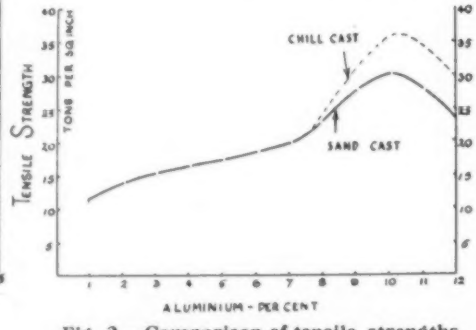

Fig. 2.—Comparison of tensile strengths of sand-cast and chill-cast aluminium bronze.

Aluminium bronze with 10% aluminium is about as susceptible to heat-treatment as steel, and the process of

1 H. Rix and H. Whitaker, *J. Inst. Metals*, 1918, vol. 19, p. 123.

die-casting confers a special type of heat-treatment by virtue of the rapid cooling which takes place. Fig. 1, based on the work of Stockdale,[2] represents the equilibrium diagram of the aluminium bronzes with up to 15% aluminium.

It will be seen that if a 10% alloy were allowed to cool slowly from its freezing point it would eventually become a mixture of α plus a eutectoid of α plus δ. As such it would approximate to the equivalent of the pearlitic condition in steels. If the same 10% alloy is moderately rapidly cooled, the eutectoid change cannot completely occur, and instead a sorbitic formation (that is, an incipient breakdown product of β) is arrived at and, as in steel, this gives greater strength than the slowly cooled pearlitic structure. As a result of this and also because the grain size is fine, due to the rapid cooling, a die-cast aluminium bronze will have a strength appreciably greater than the same alloy, sand cast.

Fig. 2 shows the tensile strength of sand-cast and chill-cast aluminium bronzes; the results are taken from the interesting book on aluminium bronze published by the Copper Development Association.[3] It will be realised that as the alloys with up to 7% aluminium are single phase, the strength will be approximately the same whether the test-bars are chill-cast or sand-cast. Above 7% the disparity between the two conditions will be observed, showing how the introduction of the second phase renders the alloy amenable to heat-treatment.

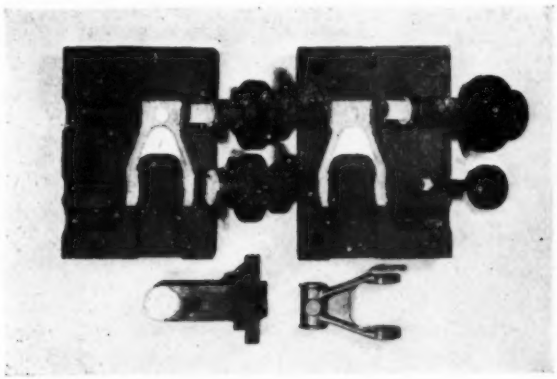

Fig. 3.—Gravity die and cores together with the aluminium-bronze component it produces.

Another characteristic which made aluminium bronze a favourite with die-casting users was its high resistance to corrosion. The alloy does take on a slight tarnish, but this prevents further attack, and so aluminium bronze has come to be widely used in conditions of severe corrosion, particularly by sea-water and certain acids. Furthermore, its resistance to fatigue under corrosion is very good. Thus D. J. McAdam[4] reports that fatigue limits (100 million reversals) of aluminium bronze, when tested in air and water, were respectively ± 14·5 tons, and ± 10·5 tons per sq. in. Other work on this subject has been done by Gough and Sopwith,[5] who use the more searching salt spray for the corroding medium. At ± 11 tons per sq. in. the aluminium bronze stood 18·5 million reversals, and at ± 10 tons, 59·3 million reversals, under salt spray.

Gravity Die-casting

Although cast-iron moulds can be used for aluminium bronze gravity die-casting, just as they were in the early days, steel dies are often preferred, because a better finish is obtained and also because of the increased life of the die. There seems to be many unexplored possibilities in die materials for gravity die-casting in this alloy, and perhaps the future will see special cast irons and steels developed for the purpose.

Fig. 3 shows a die and cores, together with the aluminium bronze component which was produced. It will be seen that some of the cores are telescopic. This is to allow the air to escape easily and thus to obtain a clean edge. In gravity die-casting, if such cores were not vented in this manner, the edges would be rounded on account of the cushioning effect of the entrapped air.

The two halves of the die are fitted together, using dowels, such as will be seen in the illustration. The cores are occasionally dipped in a dressing, such as water with graphite in suspension, and are inserted into position. Each core is made so that it can be rapidly withdrawn; in this case the cores are provided with circular collars, so that they can be extracted from the cooling die-casting by a pair of levers. Aluminium bronze has a fairly high contraction, and the removal of cores, especially large ones, is sometimes quite difficult; the core has to be pulled as soon as possible, yet not before the alloy has solidified.

When the metal is poured, experience will generally show that the tool has to be tilted to get the best results. The correct position is usually arrived at during those few trying days while the die is being run in. The inclination of the die not only affects surface appearance, but also the solidity of the casting. Sometimes it will be found that a dangerous contraction cavity can be eliminated by altering the angle of tilt of the mould.

Speed of production depends on the complexity of the part and on other factors (for example, some forms of die-casting are better poured slowly, while others need a quick feed). Small, fairly simple parts can be produced at the rate of 30 to 50 per hour, while certain difficult and large castings may only be turned out at the rate of 8 to 10 per hour. If the design is very simple, several impressions may be cut in the die and the output correspondingly increased.

The pouring temperature of the alloy is kept sufficiently high to ensure complete filling of the mould, but overheating is avoided wherever possible. The temperature of the die is maintained at between 200° and 400° C., care being taken not to get the die too hot; if the metal and die are superheated, surface appearance of the die-casting and fluidity of the liquid metal may be improved, but only at the expense of solidity. If the surface area of the part is large and the section small the metal sometimes has to be overheated to induce it to fill the die cavity, but in such parts there is a danger of getting porosity, particularly near the runner position.

The life of the mould is of the order of 10,000 castings, but cores, particularly those which are likely to become very hot owing to the proximity of the runner, will require renewing more frequently. An instance of this was experienced in a difficult aluminium bronze die-casting weighing about 3 lb. A cored hole 6 in. long and 1 in. diameter had to be included, and in spite of the fact that the core was made of alloy steel, it had to be renewed after every 500 to 1,000 castings. Indication of wear on cores is given by hair cracks in the steel and by necking at the end of the core. This gives an undersized hole with a flash across the end.

The die-castings are cooled and then inspected for surface appearance and dimensional accuracy and, of course, to see whether the part has run satisfactorily without any cold shuts. After this, the runner is cut off and all flashes removed. Some machining operations may also be carried out by the die-caster—for example, it is often found cheaper to drill a small hole than to produce it during the die-casting operation, where frequent core renewals would be necessary. On this account, holes of less than ¼ in. diameter are generally not included, because of the wear on the slender core-pins. In order to facilitate the withdrawal of cores from the rapidly contracting casting, holes are tapered by at least 0·015 in. per inch of length. For a similar reason, faces which occur parallel with the axis of the die are allowed "leave," so that the casting may be taken from the mould without scoring the surface.

The word "impossible" ought not to occur in any die-caster's vocabulary, but some design details are preferably

2 D. Stockdale, *J. Inst. Metals*, 1922, vol. 28, p. 273.
3 Copper Development Association, publication No. 31, "Aluminium Bronze."
4 D. J. McAdam, Proc. Int. Congress for Testing Materials, Amsterdam, 1927, p. 305.
5 H. J. Gough and D. G. Sopwith, *J. Inst. Metals*, 1937, vol. 60, p. 143.

omitted from the die-castings simply because they can be produced cheaper by other methods. Small holes have already been cited as an example of this. The same applies to threads, which prove unduly expensive to include in a copper alloy die-casting because of the heavy wear which would be involved in the mould. It is much more economical to cut the threads by a separate machining operation. Again, for economic reasons linked with die life, copper alloy die-castings are usually produced without undercuts being attempted.

Where machining allowance is necessary on a die-casting, it need not be so liberal as that required for a sand-casting or a steel forging (for this reason a die-casting often proves economical solely because of the raw material which is saved). On localities where machining is necessary, an allowance of $\frac{1}{32}$ in.–$\frac{3}{64}$ in. should be made. Arising out of the point about saving raw material, it may be mentioned here that the ruling section of the copper alloy die-casting can often be made less than that of the casting or forging which is being replaced. A good working average is $\frac{1}{8}$ in., and the section should be maintained as uniform as possible.

As gravity dies are not unduly expensive, and as setting-up is not such a lengthy operation as in, say, pressure die-casting, the gravity process can be used for small and medium quantities. The economic minimum depends on how many machining operations are going to be saved by using die-castings, but the process is rarely employed for quantities under 250. When numbers greater than about 10,000 are involved (calling for an output of, say, 500 per week) a second die is usually made to act as reserve in case of breakdown and to step up the output, if this should be needed at any time.

Figs. 4 and 5 show some miscellaneous aluminium bronze die-castings, the first illustration showing that quite intricate shapes can be produced, and the second illustrating some types of gear forms which can be gravity die-cast in copper alloys. Needless to say, there are many wartime applications of die-cast aluminium bronze and other copper alloys, but these cannot be illustrated here. The foregoing remarks about design, economic quantity, etc., also apply to brass gravity die-casting, which will now be discussed.

Brass Gravity Die-casting

Although aluminium bronze was comparatively simple to die-cast, brass gave difficulty owing to the volatilisation and subsequent oxidation of the zinc. This meant that in a short time the die became coated with a hard layer of zinc oxide which was most troublesome to remove. If the oxide were allowed to accumulate, the accuracy of the die-casting would suffer. In this connection the writer remembers that seven or eight years ago it was usual to specify the accuracy of brass gravity die-castings as ± 0·007 in. per inch, compared with ± 0·005 in. per inch for aluminium bronze.

There are several reasons why it became essential to perfect the gravity die-casting of brass. For one thing, it was difficult to solder aluminium bronze while, of course, brass does not present this difficulty. Furthermore, aluminium bronze has an electrical conductivity only 16% that of copper, compared with 27% for brass. Also brass is easier to machine than aluminium bronze. Then with the advent of war it became important to conserve aluminium, so brass has become preferred wherever its mechanical properties are adequate.

These various incentives were sufficient to establish the necessity of brass gravity die-castings and consequently this alloy was rapidly included as an important item in the repertoire of die-casters. Several minor problems had to be solved which consisted essentially of various changes in die design and die material, methods of pouring and venting suitable to the brass and more frequent dressing of the

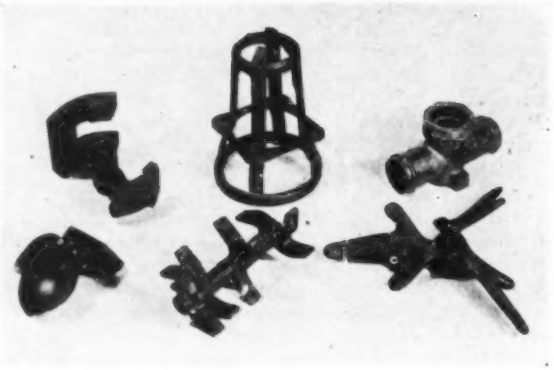

Fig. 4.—Group of aluminium-bronze gravity die-castings.

mould than had been necessary with aluminium bronze. It was an interesting case of "Needs must when the Devil drives."

Die-cast brass is based on the composition 60/40, partly because of its comparatively low melting point, and also because it is not so hot-short as 70/30. An addition of 0·2 to 0·5% of aluminium is made for deoxidation and because the brass runs better with this addition. A sign of the war-time importance of die-cast brass is found in the recently introduced emergency British Standard specification 932. Another similar emergency specification is No. 920 for Naval Brass, which includes tin between 0·5 and 1·5% and lead not more than 0·1%. It may be remarked that these specifications are purposely made generous as regards the physical properties which are expected of them. For instance, the 932 specification for brass calls for ultimate tensile strength not less than 18 tons per sq. in., and elongation on 2 in. not less than 25%. In actual practice, tensile strengths of 23–26 tons are obtained and elongation figures above 30% have been noted.

In addition to the aforementioned alloys, manganese bronze, copper-aluminium-nickel alloys, and some others have been die-cast to a rather limited extent, but doubtless during the next few years advances will be made in the subject referred to at the beginning of this article—namely, the extension of the potentialities of gravity die-casting to other copper alloys. The whole field of die-casting high-duty copper-base alloys remains invitingly open for a good deal of extra work and progress. Our American friends are almost certainly in advance of Britain in this respect. Very little shipping space would be required if they would export to us, under the Lend Lease Bill, technical information as to how they manage to die-cast, and even to pressure die-cast, high-duty copper-base alloys, which are still giving difficulty in this country.

Fig. 5.—Gear forms gravity die-cast in aluminium-bronze.

The Production of Aluminium Bronze Castings

By Frank Hudson

The successful production of aluminium bronze sand castings is still far from being a really commercial proposition, due mainly to defects promoted by the formation of oxide and dross when casting. In view of the increasing demand for aluminium bronze castings, the development of a satisfactory technique is a real need. An improved technique which involves the use of an inert mould is discussed, together with its practical applications.

MUCH has been written during the past few years on the utility of aluminium bronze as a material of engineering and, from the production view-point, the manufacture of wrought products of various kinds has been firmly established. Good results have also been obtained relative to the production of die-castings in aluminium bronze by the gravity method, but the successful manufacture of sand castings is still far from being a really commercial proposition to the majority of non-ferrous founders. This is particularly unfortunate in view of the increasing demand for aluminium bronze in the cast form, and there is no doubt that the development of a satisfactory technique would fill a very real need at the present time, and the problem is therefore worthy of some attention.

In the first place, it is clearly apparent from previous articles and papers published in the technical press that ordinary foundry methods do not lend themselves to the production of aluminium bronze sand-castings. As a result, certain methods have been advocated, entailing either bottom running or, alternatively, some modified form of the Durville process, in an attempt to avoid turbulence during pouring, commonly considered to promote defects due to oxide and dross. None of these methods has proved completely successful except for castings of relatively simple shape, and it is doubtful as to whether they ever will, in view of the design of the majority of castings required by the engineer. On the other hand, the trouble of oxide and dross formation has never been a major consideration in the production of aluminium bronze die-castings for probably two reasons :—

(*a*) The amount of oxygen (air) available inside the mould cavity is small and may even be replaced by reducing gases through the effect of mould temperature on the oil or other carbonaceous matter in the mould dressing.

(*b*) The mould material is usually steel, and therefore impermeable and inert, neither containing oxygen in the form of air, or giving rise to its formation as a reaction product.

It might then be assumed that if it is possible to pour aluminium bronze out of contact with oxygen (*i.e.*, air) into a mould which does not contain or form oxygen during the casting operation, the production of castings will be considerably simplified. In the absence of oxygen it will not matter how turbulent a casting is poured, as oxides could not be formed. It might be useful, therefore, to consider whether it is possible to obtain the above requirements in the foundry.

Production of an Inert Atmosphere

The formation of oxide during pouring can be prevented by casting in a vacuum or through the use of an inert gas. Vacuum casting not only entails rather specialised equipment, but also has an effect on other properties of the metal, such as gas content and shrinkage, and must therefore be excluded from the scope of this article. The use of an inert gas, however, appears to offer distinct possibilities providing one can be obtained having the following characteristics :—

(1) Preferably heavier than air.

Obviously, the removal of air from a mould is most easily effected by displacement with a heavier gas. Furthermore, a heavy gas is more likely to be retained within the mould cavity.

(2) Non-toxic.

(3) Non-inflammable and incapable of forming explosive mixtures with air.

(4) Stable and insoluble when in contact with molten aluminium bronze.

(5) Commercially available both as regards quantity and price.

Carbon Dioxide.—The first gas which comes to mind likely to conform to the above conditions is probably carbon dioxide (CO_2). It is 50 % heavier than air, having a specific gravity of 1·529 (assuming air to have a value of 1·0), non-toxic, non-inflammable and available in large quantities at low cost. Carbon dioxide is, however, inclined to be unstable at high temperatures, and dissociates into carbon monoxide and oxygen, the reaction being represented by the formula $2CO_2 \rightleftharpoons 2CO + O_2$. Admittedly, this reaction is slight at temperatures between 1,000° and 1,500° C., but is noticeably increased by the presence of moisture and certain metals. The production of carbon from solid carbon-dioxide by burning aluminium or magnesium in a cavity made in a block of the solid is a common and spectacular lecture demonstration experiment. Tin will also effect a reduction leading to the formation of stannic oxide along the lines of the equation :

$$Sn + 2CO_2 = SnO_2 + 2CO$$

There is some evidence to believe therefore that there is every possibility of oxide being formed on molten aluminium bronze when in contact with carbon dioxide, and this rules out the possibility of using this gas for the purpose in mind.

Nitrogen.—Nitrogen is another inert element which appears likely to conform to the required conditions. It is very slightly lighter than air, having a specific gravity of 0·967, but possesses all the other ideal characteristics with the exception of stability at temperature. Nitrogen combines slowly with lithium at room temperature to form the nitride Li_3N (more rapidly on heating), and with magnesium, calcium, strontium and barium at a red heat to form the nitrides R_3N_2. Boron and aluminium form nitrides BN and AlN at a bright red heat. It is evident, therefore, that whilst the use of nitrogen will prevent the formation of oxide on molten aluminium bronze, it may lead to the presence of nitrides, and just what effect this may have on the properties of the final casting without experimental evidence being available is hard to say. Maxwell[1] has reported some interesting data on the action of nitrogen upon aluminium bronze. Using 3 : 1 hydrogen-nitrogen gas mixtures at 1,000 atmospheres pressure, temperatures up to 500° C. and times up to 1,535 hours, a deterioration of mechanical properties and a depletion of aluminium in the surface layers was experienced. This was found to be due to an interaction between nitrogen and aluminium, and it would thus appear that aluminium bronze should be used with caution under conditions where such an interaction could occur. So far as present considerations are concerned, the use of aluminium bronze is essential, so the

1 H. L. Maxwell, *Trans. Amer. Soc. Metals*, 1936, vol. 24, p. 213.

caution mentioned above must be considered in the reverse direction and the use of nitrogen viewed with suspicion.

Argon.—The third and only other inert gas which appears to meet requirements is argon. This gas has a specific gravity of 1·38 and is in demand for filling electric lamp bulbs. It is stable, so far as we know, in the presence of all molten metals, and is not absorbed by the metal. Bulk supplies are available, and although the price of this gas is much higher than either carbon dioxide or nitrogen, it is not too high to prevent practical application. Furthermore, in view of its use in electric lamp bulbs, it will obviously prevent the formation of oxide, and undoubtedly provides a solution to the first consideration relative to a more positive method of producing aluminium bronze castings—namely, the production of a suitable inert atmosphere.

Production of an Inert Mould

The ordinary sand mould, whether green or dry, provides a plentiful supply of oxygen. Not only are the minute voids between the sand grains filled with air, but appreciable quantities of moisture are present which is readily decomposed on contact with molten metal into hydrogen and oxygen. To prevent the formation of oxide on molten aluminium bronze after it has entered the mould, it is obvious that the latter must be rendered inert so far as the formation of oxygen is concerned. This can probably best be arranged by using an impermeable facing sand to reduce the available pore space holding air to the lowest limit, and then, after thorough drying, to spray the mould surface with a fairly heavy layer of some metal by means of the metal pistol. The metal selected for spraying the mould surface should, of course, have a melting-point well in excess of aluminium bronze, and the use of pure iron or nickel might be considered for this purpose. This sprayed coating should act in two directions, sealing the surface of the mould so that the inert gas used for filling the mould is retained in position, and, secondly, preventing access of oxygen from the mould material to the metal being poured. A coating of soot, deposited by means of the acetylene flame, on top of the sprayed metal coating, might be required to give added protection against molten metal attack. Special attention will of course have to be given to those parts of the mould where metal impingement occurs, such as in the neighbourhood of ingates, and it will be a wise precaution to prevent overheating of the sprayed mould coating at such points by the use of metal or carborundum chills. Moulds made from high conductivity moulding compositions,[2] such as carborundum in conjunction with a metal-sprayed surface should give excellent results. The rapid dissipation of heat would permit the use of lower melting-point metals, such as copper, for surfacing the mould and reduce the time factor, in which any gas reactions likely to form oxide could take place. Some difficulty will undoubtedly be experienced in metal spraying the inside diameter of small down-runners, but this can be overcome by making the runners in the form of split inserts or by the use of impermeable refractory tubes. It should, of course, not be forgotten that if oxide formation can be prevented the need for bottom-pouring methods entailing the presence of down-runners ceases to be of major importance.

So far as cores are concerned, oil-sand could be used if desired, providing these are metal sprayed in the manner already outlined.

Practical Applications

To illustrate more clearly the application of these new ideas to the manufacture of aluminium bronze castings, it might be of interest to give a practical example for, say, the mass production of small marine propellers, a service for which the alloy is particularly useful.

Fig. 1 illustrates a method which should be capable of application in any foundry. The mould is constructed in the normal manner with the exception that the usual down-runner invariably employed for bottom-pouring such castings in aluminium bronze is omitted. A cast-iron chill or piece of carborundum brick is inserted in the mould at A to prevent overheating of the mould at this point, due to metal impingement from the combined pouring basin and runner B. Care is taken to carry ample vents between $\frac{1}{8}$ in. to $\frac{1}{4}$ in. in diameter from the top edge of the blades to ensure that the gas inside the mould can escape when pouring commences. After drying, the whole surface of the mould, including riser, is metal-sprayed as previously described, taking care not to close up the vents in the operation. Joints between moulding boxes must be sealed externally with a putty-like paste of linseed oil and clay.

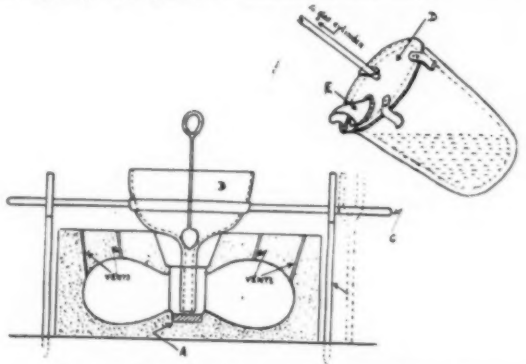

Fig. 1.—Proposed layout for casting aluminium-bronze in an inert atmosphere to prevent oxide formation.

The combined runner and pouring basin B is made in cast iron and is equipped with an accurately fitting ground-in plug for controlling the supply of molten metal to the mould. The complete pouring basin assembly is well heated before use, and is capable of being lifted by the usual double-handled crucible carrier C.

A lid D, made from heavy gauge steel or cast iron, will be required to fit over the crucible used for melting. This is kept in place by external clips, and is fitted with a length of steel tubing for connecting to inert gas cylinder. This lid is also fitted with a gas deflector shield, as shown at E over the pouring orifice.

The aluminium bronze required for casting should be melted in an ordinary lift-out crucible furnace under a suitable slag for removing alumina—there are several of these on the market. The pouring of molten aluminium bronze through air as from a large furnace into carrying crucibles or shanks must be studiously avoided, as this is exactly what we are trying to prevent. Excessive agitation by stirring should also be forbidden. When the crucible is lifted from the furnace the slag should be removed by careful and gentle skimming and the lid D fitted to the crucible and enough argon admitted to completely *fill* the space between the top of the molten metal and the lid. The amount of argon required can be gauged by means of a lighted taper held in front of the pouring orifice in the lid. When the taper is extinguished the space inside the crucible is full of inert gas. At this stage it should be noted that it is obviously necessary to have a fair amount of argon inside the crucible and therefore the crucible should not be too full of metal. The ratio of metal weight to crucible size should be taken into account beforehand. The crucible, still connected to the gas cylinder, but with the argon turned off, is now carefully carried, without splashing, to the mould for pouring.

Coincident with the metal being lifted from the furnace the heated cast-iron pouring basin and runner B, supported on trestles, has been fitted over the mould, so that it is just touching the top of the mould. The plug is removed to one side and the mould filled with argon from a second cylinder by means of a metal tube introduced down the runner. The inert gas should be admitted slowly to enable the air inside the mould to be completely displaced by means of the vents. The completion of this operation can be confirmed

(*Continued on page* 193.)

2 F. Hudson, *Metal Industry*, May 21, 1937, p. 475

Increasing the Efficiency of Iron and Steel Works

The importance of systematic work- and time-studies, with a view to the improvement of efficiency in a works, cannot be overestimated; they so often bring to light many sources of waste, many of which are quite obvious but are habitually overlooked. Some results of investigations are given which have been applied with great success.

AMAZING changes have been made in the management of industrial organisations since the early days when the chief executive was brought into close contact with his employees. The trend towards larger units, the demand for better products at lower prices, and the general increased rate of production by the greater application of machinery have introduced more complex problems in management, with their resultant difficulties in maintaining operation on efficient lines. Fundamentally, management, i.e., the handling, controlling and utilising to the full the abilities and efforts of those employed in a particular works, is an art, although scientific methods are being increasingly employed to improve the efficiency of the works and at the same time to ensure for the employees a fair return for their work.

The iron and steel industry is an example of an industry which is now largely carried on by big units, and unless great care is taken wasteful managerial methods will greatly impair the efficiency of an organisation. This is true of normal peace-time activities, but under conditions of war, when there is urgent need for maximum speed in production, there is a tendency for waste in one form or another to increase, and it is necessary to keep a careful check on operations to ensure that the most efficient methods are employed. Many aspects of these modern methods for co-ordinating economic, technical and psychological factors were discussed in earlier issues* in an effort to contribute to the development of an improved industry. But it may be of interest to become acquainted with further examples of the application of results from special investigations† on work- and time-studies in an iron and steel works which have led to increased efficiency in the present emergency.

Discharging Raw Materials at the Docks

Much discontent was experienced among the men working at the docks, due to difference in earnings, and it was felt that an investigation was necessary to get at the root of the trouble and, if possible, to simplify the method of payment. The men at the docks belonged to different departments, and the method of payment had grown with the department to which the men were attached; thus, for instance, one section had only the usual shift time, while another section, concerned mainly with maintenance work, had to work much overtime. The former were paid at so much per ton of materials discharged, while the latter remained on time rates. The whole of the work had become specialised, and the relation and extent of the work had changed considerably.

Detailed work-and-time studies resulted in defining standards for the work involved—thus, the discharging of material was subdivided into: (1) from barge direct to wagon; (2) from barge via crusher to the wagon; (3) from barge to storage; (4) from storage to wagon; (5) from storage via crusher to the wagon. Further, it was found that some of the maintenance work could be done during the shift, when stoppage of the work proper could not be avoided.

The investigations resulted in piece-work wages being developed for all the men in which the total working time of the group of men working at the docks was used as a basis. Thus, a working community of the men was created—all interested in saving time where possible. Equivalent figures of the discharged tons were used in order to allow the addition of times for various kinds of work, and a value termed *Standard ton per working hour* was developed as the basis of payment. The necessary maintenance work was entered in "a repair book" with all details. A great part of this maintenance work could be carried out during times when discharging was stopped for other reasons.

When these measures were applied the need for overtime practically disappeared, the number of men employed were reduced to 90% of the former total—the remaining men being transferred to other work—and an increase in output of about 15% resulted. A corresponding increase of earnings of the workmen took place and discontent ceased.

Discharging Material by Tilting Wagons

The principal factors or variables influencing the time of discharging material by tilting wagons was investigated in a somewhat similar manner to that already mentioned. The time involved was found to be dependent upon (1) the arrival of the wagons at the tilting mechanism, which were often delayed by insufficient capacity of the connecting rails; (2) the removal of the empty wagons; (3) the weather. It was found that these unfavourable circumstances had led to an intolerable increase in demurrage month by month.

The average number of wagons per shift previously discharged was made the basic performance, and a premium paid for each wagon discharged in excess of this number. This resulted in an increase of about 20% in the number discharged, a considerable increase in the earnings of the men, but also in a decrease of demurrage costs, which amounted to about fifteen times the additional amount paid in wages.

Scrap Shears at a Strip Mill

It was considered desirable to develop a suitable piece-work system for shearing the ends of strip at a strip rolling mill. The length of ends varied from about 1 ft. to 16 ft., and the thickness from about 0·008 in. to 0·04 in. It would have been practically impossible to apply individual piece-work prices as too many different jobs had to be done, and a more superficial, but still sufficiently accurate method, had to be applied. An investigation showed that an output of 500 lb. per hour on an average could be obtained with average dimensions of the strip ends, but it was regarded as desirable that an increase in earnings should take effect from the start, and an output of 400 lb. per hour was used as a basis. This basis was permitted to operate up to an output of 700 lb. per hour, beyond which the excess was calculated on a basis of 550 lb. per hour.

This method means a higher payment to the men for outputs between 400 and 700 lb. per hour, and a lower relative payment for output in excess of 700 lb. per hour, and the results in practice proved favourable. The average output increased to 750 lb. per hour, with peak outputs of 925 lb. per hour.

Avoiding Lost Time at a Plate Mill

When carrying out investigations at a plate mill, with a view to the introduction of a piece-work system, it was found that unnecessary stoppages of production caused lost time of 20% to 40% of the working time. It was

* F. L. Meyenberg, "Industrial Management and Production Control," METALLURGIA, Oct., 1938, to Jan., 1940.
† *Archiv für das Eisenhüttenwesen*, 13, 6, 1939.

considered that these figures could be lowered to about 10% by an improvement in the flow of work on the application of the following suggestions: (1) Waiting time for trucks should be entirely avoided by keeping a truck always available at the disposal of the dischargers; (2) maintaining at hand a supply of necessary accessories such as wedges, wood, wire, hoop iron, etc.; (3) delivering guard-plates for edges to the discharging gang ready cut; (4) providing adequate light over the whole working area; (5) tare weight of the conveyer barrels, which varied between 11 lb. and 13 lb., to be written on each of them; (6) maintaining adequate space for storage of the plates and to enable the dischargers and shearmen to work side by side; (7) when delivery of plates is by lorry, they should be piled up on timber convenient for easy handling by the crane; (8) piles should be well packed, and lots not too heavy; (9) since the plates are counted at the straightener and shears, it is unnecessary to count them when loading; (10) leather handling pads should be delivered to the dischargers cut to the right size.

Loading at a Plate Mill

In connection with the above mill the workmen expressed a desire for piece-work wages for packing and loading the plates, because the premium bonus system in use for this purpose had not been a success. After the investigation above, a table of standard times was developed upon which payment was based. This table is given in the following:—

STIPULATED TIME FOR PILING, PACKING AND BUNDLING PLATES 2 FT. 8 IN. TO 8 FT. LONG AND 0·016 IN. TO 0·12 IN. THICK.

Piling plates into wagons	20 mins. per ton.				
Packing the plates—					
Number of plates per bundle	2	3	4	5	6
Time in minutes per bundle	38	42	46	50	54
Counting the plates	1·5 mins. per 100 pieces.				
Bundling the plates—					
Number of plates per bundle	2	3	4		
Time in minutes per bundle	8·0	9·6	11·2		
Putting up the plates	5 mins. for once.				

Before applying the results of these investigations 12 men were employed on the loading and packing. Afterwards the same amount of work was accomplished by four men, upon which only half of their time was occupied, and by two men who were doing work for payment at an hourly fixed rate.

Planing Machines in a Machine Shop

An investigation on the work accomplished by some planing machines showed that the feed and speed of these machines did not meet the demands which could be put upon modern tool steel. It was discovered that when the tools were ground to more economic cutting angles, and the power increased, that the cutting times could be greatly augmented. The feed, speed and the cross-section of the chip produced could be satisfactorily increased up to 50%, with a consequent greatly increased production.

Work in Laboratories

Overtime had become the rule in certain laboratories and it was decided to examine the cause and find out whether it was really necessary and how it could be limited to a tolerable extent. The monthly demands of the various departments for the variety and the number of tests and analyses were collected in statistical tables. These were discussed with the managers of the departments concerned, and it was finally decided that the following analyses were not necessary:—

Routine analysis for the blast furnace, 5%; for the Bessemer plant, 15%; for the open-hearth furnace plant, 10%; while for other departments an average of 5% could be saved. This amounted to about 400 hours working time. A second means of reducing pressure upon the laboratory staff was to adopt some means of distinguishing the various casts of the melting plants in order that the steel which required to be of accurate composition could be specially indicated, while for other casts greater tolerances were allowed. Thus the production departments were able to decide whether it was desirable to repeat an analysis when the steel was specially indicated. Before these investigations it had been customary to duplicate all analyses.

Standardisation of Wire Ropes

An investigation at some works indicated that only 57% of the wire ropes used corresponded to the national standardisation requirements, while the remaining 43% were ordered by the departments concerned more or less at random. Deviations from the national standard were considered and, where reasonable, were retained for use, but the result of the investigation effected a reduction in the number of diameters of the wire ropes from 51 to 24, i.e., to 47%, and of the various makes from 25 to 4, i.e., to 16% of the original stock. A corresponding decrease was also obtained with regard to the sizes of the wire used in the manufacture of the ropes used. The advantages of standardisation are well known, but it may be mentioned, in this case, that better use of the wire-rope manufacturing machines, due to less changes from one make to another, is one advantage, and less stocks need be carried.

Accurate Work Instructions Facilitate Piece-Work

The introduction of a new piece-work rate is sometimes difficult, but it can be facilitated by accurate instructions on how the work should proceed, such as adopted in one works in connection with the colouring of reels for barbed wire. Many different colours were used, and the reels were put on rods and dipped into tubs filled with the particular colour desired. After the reels had been allowed to drain they were transported by carts and piled up in the store. The instructions for this work were as follows:—

(1) The colour must be put in the tubs by the piler, not by the painter. During the filling of the tubs the painting must continue.

(2) When changing carts the painter to take care that an empty cart is at hand immediately after the departure of the full cart so that the crane has not to wait.

(3) The reels to be arranged so that six can be taken up at one time by the crane. The painter should arrange the uncoloured reels during the travelling of the crane to the tubs, the dipping and draining, and the piler has sufficient time to help him.

(4) The piler and not the painter should return the rods of the reels to the stores. He should also exchange damaged parts of the reels and clear the workplace so that the work of painting does not stop.

(5) The colour must be stirred continuously including meal-times, and when the cranes are needed for other work. For this purpose a mechanical stirring device is set in operation at the beginning of these stoppages.

The introduction of piece-work rate in this case resulted in an increase of output of 40%, and it is probable that the detailed instructions were one of the main reasons for this success.

It has only been possible to summarise a few of the results of investigations which have been successfully applied. In their discussion attention was directed to some points of general importance which are of particular interest, and can be briefly referred to here. In some cases it is shown that the formation of working groups, for a rate covering the complete operation, is preferable to the use of the individual piece-work rate system. In the former method the working community is built up of men doing different work, but the final result of the work is dependent upon each, and it is distinct from an individual man or a group of men doing the same work. It is emphasised that the importance of systematic work-and time-studies cannot be overestimated; they so often bring to light sources of waste, any one of which may be regarded as quite obvious, but which is habitually overlooked.

Testing Aluminium Alloys

Need for Standardisation of Fatigue Testing Methods

By H. F. James

Director, Northern Aluminium Co., Ltd.

A considerable measure of standardisation in the testing of aluminium alloys has been introduced, particularly in the case of those materials which conform to official specifications, but these do not apply to the fatigue testing of these alloys, nor to the measurement of their strength at high temperatures; the views of the author on this subject are, therefore, of special interest.

AN urgent need exists for the standardisation of the methods employed in the fatigue testing of aluminium alloys, and in the testing of their mechanical strength at elevated temperatures. In assessing the merits of various materials for a particular purpose, designers usually have available a wealth of quantitative data regarding the mechanical properties of each material. In cases which involve only simple designs, or which merely entail modification of a design on which long experience has previously been gained, the data available is usually sufficient to enable a sound choice of materials to be made. By applying a reasonable factor of safety, it can be ensured that, under conditions of static loading, the limiting stress which the final product will be required to withstand will lie well within the mechanical properties of the material.

Into the general testing of metal and alloys—i.e., testing for ultimate tensile strength, proof stress, elasticity and hardness, a considerable measure of standardisation has been introduced, particularly in the case of materials which conform to official specification, such as those issued by the British Standards Institute and the Directorate of Technical Development. The data obtained from such tests, therefore, enables accurate comparisons of materials to be made.

Unfortunately, this does not apply to the fatigue testing of aluminium alloys, nor to the measurement of strength at high temperatures, although it is in these two particular directions that the most accurate information is necessary.

It is widely known that any metal or alloy will fail under a stress considerably lower than its ultimate stress if the load be applied and removed a large number of times. This applies to all types of stressing actions—i.e., axial loading, bending and torsion—and cases where a vibratory or alternating stress is superimposed on a static stress. If, therefore, failure under such alternating stresses is to be guarded against, accurate information must be available regarding :—

(*a*) The number of stress cycles necessary for a given stress, to cause failure ; and

(*b*) The limiting stress range, if any, for the material below which failure will be unlikely to occur no matter how long the duration of the test.

Several types of machines are used for carrying out fatigue tests, the most common being that known as the rotating beam machine. The types of tests most frequently applied are those involving complete reversal of tension-compression stresses, but as will be readily appreciated, many different combinations of stressing and straining actions are possible.

In view of the complexity of the stresses which may actually be encountered, it is desirable that the conditions of the test should be arranged to simulate, as closely as possible, those to which the actual component will be subjected in service. It is, however, universally admitted that in the majority of fatigue tests the conditions of testing are far from conforming to this requirement.

For example, tests involving tension-compression stress reversals are only directly applicable to cases where the material under test is to be used for components which will be subjected to such a type of stressing action. This is rarely found to occur in practice, although Hartmann* quotes a railway car axle, which rotates in its journal so that the extreme fibres pass alternately through tension and compression, as an example where such a test may justifiably be employed.

In most cases, stresses encountered in actual operation are so complex as to render it almost impossible to reproduce them accurately in a test, this naturally reducing the value of the test result. Also, the issue is further complicated by the fact that the result of the test is greatly influenced by the shape, size and surface finish of the test-piece and the physical condition of the material.

Thus, it is not surprising to find that a particular material, such as, for example, heat-treated duralumin, when subjected to fatigue tests employing different stressing and straining actions, will give consistent results under each set of conditions, but will show appreciable variation of results from one set of conditions to another.

In some cases the possibilities of error have been minimised by resorting to full-scale testing, that carried out on certain members of the Sydney Harbour Bridge furnishing a classic example. For obvious reasons, however, it is impossible to adopt this practice in every case, and thus the need for reliable small-scale test data remains.

A vast quantity of work has been carried out to meet this need, but in attempting to overcome the inherent difficulty of the problem, types and conditions of testing have been so varied that the present position, from the designer's point of view, is one of more or less complete chaos. It is, therefore, becoming widely felt that methods and equipment for fatigue testing should be subjected to a considerable measure of standardisation, so that test results will enable at least reasonably accurate comparisons to be made of different materials.

Results of standardised tests would provide designers with accurate knowledge of the behaviour of materials under standard conditions regarding shape, size and surface finish of test-piece, physical condition, and type of stressing and straining action, and should at least provide a basis from which to work in any particular job. Literature on the subject could be directed into well-specified channels, and would then undoubtedly be of considerable help in design, instead of providing, as at present, a maze from which no one can hope to escape except in a state of utter confusion.

The testing of a material at elevated temperatures does not involve the intricacy and difficulty inherent in fatigue testing. The problem is, in fact, closely comparable with that encountered in the testing of aircraft performance. In this latter case, as is well known, figures obtained in actual tests are not themselves officially accepted, they are merely used to compute the performance in an atmosphere standardised with respect to all variables for each altitude. Comparisons of the performances of different aircraft are, therefore, highly accurate.

Similarly, if comparisons of mechanical strength of aluminium alloys at elevated temperatures are to be accurate, the test data should be such as to obviate any

* "Fatigue Test Results." By E. C. Hartmann. *Product Engineering*, February, 1941.

factors due to differences in testing procedure and conditions. The simplest way to ensure this is to standardise the tests with regard to size, shape and finish of test-pieces, temperatures of testing, time of soaking, and rate of straining.

It is also important that the test conditions should approximate as closely as possible to those encountered in service. Thus, in the testing of piston materials, it is particularly important that soaking be carried out for an adequate period. If the soaking period is too short, considerable temperature differences will exist between the surface and the centre of the test-piece, and the result obtained will in no way be indicative of the strength of the piston after some hours of running in an internal combustion engine.

Lack of such standardisation, and the failure to state the conditions of testing when publishing the results of tests, has led to considerable confusion and to the making of valueless comparisons of the behaviour of the various light alloys available for use at elevated temperatures. Until such time as the necessary measure of standardisation is introduced designers should, when making a choice of material, take close account of the conditions of testing, particularly with regard to time of soaking and rate of straining.

Sheffield Society of Engineers and Metallurgists

A JOINT meeting of the above Societies was held in Sheffield on October 4, to discuss four papers recently submitted to the Iron and Steel Institute. Members of the Institute were also invited and the meeting, which was presided over by Dr. W. H. Hatfield, F.R.S., was largely attended. The papers presented for discussion included "Examination of High-Sulphur Free-Cutting Steel Ingot," by E. Gregory and J. H. Whiteley; "The Thermal Relations Between Ingot and Mould," by T. F. Russell; "The Formation and Properties of Martensite on the Surface of Wire Rope," by E. M. Trent; and "Note on the Resistance to Furnace Atmospheres of Heat-Resisting Steels," by A. G. Quarrell. The main part of the discussion is given in the following notes:—

Examination of High-Sulphur Free-cutting Steel Ingot[1]

Mr. N. H. Bacon, of Steel, Peech and Tozer, Ltd., expressed his interest in the fact that high-sulphur steel rolled in such a manner without cracking. Some years ago his firm found that as the sulphur content in steel increased from 0·03 to 0·05 or 0·06% cracking increased, but a few days ago the firm ran a cast containing 0·35% sulphur and it rolled quite well. It was also found recently that a steel containing 0·2% sulphur with 0·2% lead cracked much less than a 0·2% sulphur steel without lead, and it seemed to him that lead had an effect similar to an increase in sulphur.

Prof. J. H. Andrew, of the University of Sheffield, said one thing struck him most forcibly: it was that sulphur appeared to be an excellent deoxidising agent. This paper seemed to give further evidence in favour of the theory, proposed some years ago by Dr. Trent and himself, that the segregation in an ordinary ingot was due to the CO developed by the reaction between FeO and C. When they added sulphur the reaction stopped. In his opinion, instead of getting oxygen in a gaseous combination in a high-sulphur steel, it was used up to form slag, and for that reason normal segregation was absent in high-sulphur steel.

Dr. Hatfield referred to a major research he and his colleagues had been carrying on for some years, dealing with the effect of varying amounts of work in various steels, in connection with which he took half of the ingot described in this paper, hammered it down, and put varying degrees of work upon it, and then submitted it to longitudinal and transverse tests. The resulting figures were shown on a slide. The material responded in a remarkable degree as a homogeneous material, and the figures showed remarkable ductility, having in mind the high sulphur content; he said the Park Gate Co. were to be congratulated on the properties displayed by this type of free-machining steel.

Mr. W. W. Stevenson, of United Steel Companies, Ltd., referring to Dr. Gregory's remark that "at the moment there appears to be no absolutely reliable method of analysing the inclusions," said they could isolate them by the alcoholic iodine method and by fractional vacuum fusion they could also indicate the form of occurrence of the oxides. In the ingot described, the only element that segregated to any great extent was oxygen, and it seemed a pity no cast history was available which would indicate the oxygen condition from charging to tapping and enable them to determine what happened during the heat. He suggested that, if samples were available, some attempt should be made to obtain the constitution of the oxygen. On being advised by Dr. Hatfield that samples were available, Mr. Stevenson, on behalf of Dr. Swinden, said they would be glad to examine and report upon them.

The authors promised to reply to the discussion in writing.

The Thermal Relations Between Ingot and Mould[2]

Mr. T. Land, of William Jessop and Sons, Ltd., said he thought Mr. Russell had shown that a fair degree of correspondence could be obtained between the actual temperature relations in a mould and the theoretical relationship which could be calculated by mathematical methods. This paper constitutes a definite contribution to the theory of ingot mould design.

Mr. J. Woolman, of Brown-Firth Laboratories, said this work by Mr. Russell was of great value to those who were concerned with heat transference in large masses, particularly in ingot moulds, but he failed to see how it could apply to the calculation of stresses. For the size of ingot concerned, it had been shown that there was no advantage in having an ingot mould ratio in excess of 1; it would be interesting to know, however, whether that ratio applied to other sizes of moulds. Did it apply to a very large mould or a very small one? The question of the separation of the ingot from its mould introduced a serious error in the calculation for temperature distribution, and the differences between theory and practice, shown in Mr. Russell's curves, were partly due to that fact.

The Formation and Properties of Martensite on the Surface of Wire Rope[3]

Mr. J. H. Whiteley stated that occasionally he had seen a small amount of martensite in the head of an ordinary steel rail, brought about by the friction of the wheels on the rail. With regard to Dr. Trent's statement that "a coarser structure would require either a longer time or a higher temperature for the complete transformation to austenite," Mr. Whiteley thought experiment would show there was very little difference between the speeds at which relatively coarse and very fine structures disappeared. He did not find any valid evidence for the statement that nitrogen was not present in the hard layers.

Mr. P. Woodhead considered it misleading to say that wire was used in a sorbitic condition. Heat-treated and tempered wire was no good for mining ropes. With regard to the flowed skin, he was not altogether in agreement that it was martensite; he thought it was an amorphous skin that was formed by the friction.

Contributors to the discussion on the paper[4] by Mr. A. G. Quarrell were Mr. H. J. Goldschmidt, of William Jessop and Sons, Ltd., and Dr. A. H. Jay, of United Steel Companies, Ltd.

1 Iron and Steel Institute. Advance copies issued June, 1941. See July, 1941, METALLURGIA, p. 85.

2 Iron and Steel Institute. Paper No. 4/1940 of the Committee on the Heterogeneity of Steel Ingots. Advance copies issued December, 1940. See January, 1941, METALLURGIA, p. 84.

3 Iron and Steel Institute. Advance copies issued March, 1941. See April, 1941, METALLURGIA, p. 171.

4 Paper No. 2/1941 of the Alloy Steels Research Committee, Iron and Steel Institute. Advance copies issued May, 1941. See June, 1941, METALLURGIA, p. 47.

The Polishing of Cast-Iron Micro-Specimens and the Metallography of Graphite Flakes*

The Metallography of Inclusions in Cast-Iron and Pig Irons†

THE above papers, which were originally presented by Mr. H. Morrogh at the May, 1941, annual meeting of the Iron and Steel Institute, were presented by the author at a joint meeting of the Manchester Metallurgical Society with members of the Iron and Steel Institute, held in Manchester on October 8, at which members of the Institute of British Foundrymen were also invited. The work described in these papers, and the manner in which it was discussed by Mr. Morrogh at this meeting, were much appreciated by an interested audience.

In summarising the work reported in the two papers, Mr. Morrogh stressed the importance of care in preparing cast-iron specimens for micro-examination, and stated that, due to inferior polishing methods, the graphite flakes were removed and photomicrographs reproduced not the graphite as many imagined, but the hole where the graphite had been ; for this reason the graphite flakes had invariably been shown black. As a result of the development of a new technique in polishing, which is described in one of the papers, it has been possible to prepare micro-specimens of cast iron without disturbing the graphite flakes. The aim achieved by his method is a smoothly polished metallic matrix free from surface distortions and smoothly polished graphite flakes, with no burnishing or flowing of the metallic matrix over the graphite flakes. The process enables the complete preservation of the graphite flakes and the polishing operation, as is indicated by the micrographs shown in the papers and by lantern slides, produces specimens perfectly prepared for photomicrography. When carefully prepared specimens of cast iron are examined at low magnification, the graphite flakes appear to be light grey to brown in colour, and any defect on the surface of the graphite gives it a black appearance.

Using the method of polishing which has been developed, Mr. Morrogh has shown in his second paper that it is possible to observe the internal structure of temper carbon nodules in malleable cast iron. The author dealt with the structure of temper carbon nodules from the point of view of the effect of iron and manganese sulphide on the type of temper carbon formed. In malleable irons containing iron sulphide —for instance, whiteheart malleable irons—the temper carbon is in spherulitic form. The basal planes of the graphite crystals, of which the temper carbon nodule is composed, are at right-angles to the radii of the spheroid of which the nodule can be said to be composed, and crystals of iron sulphide in the centre of some spherulitic temper carbon nodules were shown by means of a lantern slide. In malleable irons which contain all the sulphur in the form of manganese sulphide, the temper carbon appears to be an aggregate of very small graphite flakes having no particular orientation with respect to each other.

Although time did not permit the author to deal with the subject of inclusions in cast iron very thoroughly, his paper indicates the importance of sulphur compounds in determining the structural characteristics of malleable cast iron, and the importance of these constituents in ordinary cast iron may well be of the same order.

A very interesting discussion followed, in which many members of the Society took part, but unfortunately no metallurgist present seemed to have expert knowledge on cast iron, and although all were appreciative of the work done by Mr. Morrogh, the writer is in some doubt as to whether the significance of his work was really understood. This work has provided facilities for acquiring improved knowledge on cast irons and can be regarded as an important contribution to the metallurgy of these complex alloys.

*See METALLURGIA, March, 1941. p. 151.
†See METALLURGIA, December, 1940, p. 41.

Molybdenum

Two booklets have recently been published which are full of information on molybdenum. One presents the physical properties of molybdenum and the figures given are supported by a bibliography containing 81 references. All physical properties are given, including wave lengths of the principal lines in the emission spectra. The second booklet is concerned solely with the fundamental effects of molybdenum in steel which have led to the successful use of the many molybdenum steels that are now available. Reference is made to the industrial forms of molybdenum and particular attention is directed to its effect on mechanical properties, thus its effect on the tensile properties of steel, on hardness, fatigue, impact strength, elevated temperature strength, wear resistance and corrosion resistance, are all briefly discussed in the light of research and experience. Much attention is also given to the fabrication and heat-treatment of molybdenum steels and to the effects on the transformation characteristics when molybdenum is added to plain carbon and alloy steels in progressively higher percentages. It is noteworthy that most molybdenum steels are free-sealing during forging operations : thus forgings have good surfaces, die life is prolonged, and pickling costs are reduced. The forging procedure does not differ materially from the practice employed in the case of other alloy steels. Reference is made to case-harding and nitriding ; machining ; welding ; flame hardening ; and the effect of molybdenum in steel on these operations. This second booklet is essentially of a practical character.

These booklets are entitled " Physical Properties of Pure Molybdenum " and " Molybdenum : Fundamental Effects in Steel," respectively, and copies of either may be obtained free from Climax Molybdenum Co. of Europe, Ltd., 2-3, Crosby Square, London, E.C. 3. Application should be made on business notepaper, stating the particular booklet desired.

The combined August-September issue of the *Nickel Bulletin* contains a number of unusually interesting abstracts. Copies may be obtained, free of charge, on application to the Mond Nickel Co., Ltd., Grosvenor House, Park Lane, London, W. 1. Individuals are asked to state the nature of their interest.

The Production of Aluminium Bronze Castings

(*Continued from page* 188.)

by again holding a lighted taper over one of the vents taken from a position coincident with the highest point on the casting. The tube for admitting argon can now be withdrawn from the mould and allowed to fill the pouring basin, refitting the plug before the basin is completely full of gas. The gas cylinder used for filling the mould can now be put on one side and the mould poured.

The crucible containing the molten aluminium bronze should be gently lifted, so that its lip just rests over the edge of the metal basin, and a slow flow of argon should be recommenced as the metal issues from the pouring orifice in the lid. The transference of the metal to the pouring basin should be gentle and the metal flow should at no time strike the gas deflector. The purpose of this deflector is to blanket the metal with argon as it leaves the crucible. Just enough metal should be poured into the basin as will fill the casting and the supply of inert gas to the crucible should not be stopped until the required amount of metal is obtained. The plug can now be lifted, and as the molten metal rises in the mould the pouring basin and runner is also raised accordingly, so that when the basin is empty the runner is just leaving the still molten metal in the mould riser.

In conclusion, one can quite well imagine those metallurgists and practical foundrymen who have taken the trouble to read this article asking themselves the question—Will it work ? Well, why not try it and see !

Intergranular Corrosion in Austenitic Stainless Steels

(*Continued from page* 174.)

—e.g., between dies in a large press—the whole of the plate being heated for the process to the proper heat treatment temperature of the steel and cooled freely in the air immediately it is completed, it is probable that subsequent heat treatment would not be essential in a steel which possessed a "safe period" of 15 or 30 mins. at 650° C., though it may be noted that heat treatment might be advisable in order to release internal stresses set up in the steel should the latter have cooled below a good red heat while the dishing or flanging operation was in progress. If, however, the operation was done in sections, involving perhaps half a dozen or more local heatings, subsequent heat treatment would almost certainly be advisable both on the grounds of restoring to the steel its best structural condition and also of removing the internal stresses which would undoubtedly be set up in the steel by working it in such a piecemeal fashion.

The production of equipment which will be heated in the sensitizing range for prolonged periods during use and concurrently exposed to corrosive action, presents special problems. Complete freedom from intergranular attack is the goal for such equipment, and at first sight it might appear to be achieved by using steels containing adequate amounts of carbide-forming metals and giving the plates, or other forms in which the material will be used, a stabilising treatment at 850°–900° C. before fabrication. Such stabilised material is certainly free from intergranular attack when subsequently heated for long or short periods in the sensitizing range, but it only retains this property if it is not heated, after stabilising, to higher temperatures than were used for that operation. One sees many references in American technical literature to the merits of a stabilising treatment applied to plates, sheets, etc., of steels containing titanium or columbium, before they are fabricated into equipment, but little attention is given to the fact that welding operations remove wholly or partly from the material in the neighbourhood of the weld, the benefits conferred by stabilising. That this is the case is evident from the fact that these areas are heated during welding to temperatures approaching the melting point of the steel, and hence, as has been pointed out in an earlier part of this article, will actually develop susceptibility, during subsequent heating in the sensitizing range, more rapidly than the normally softened steel, let alone the stabilised material. The actual weld metal may, and very probably will, be in a still worse condition because part of its content of stabilising alloy—whether that be titanium or columbium—will have been oxidised during melting down.

If a stabilising treatment is essential, the logical stage at which it should be applied is to the completely welded vessel in order that the parts most in need of it—i.e., the welded joints—may benefit. It is agreed, of course, that treatment of a large welded vessel is likely to be a more difficult and more expensive operation than that of the individual plates from which it is constructed—it may even be impracticable,—but it would seem rather futile to stabilise the plates prior to fabrication and then ignore the greater need of the welded joints. The opinion has been expressed in America that the situation can be met by using stabilised titanium-containing plates and welding with electrodes containing columbium to the extent of not less than ten times their carbon content. Even if one assumes that little columbium is oxidised during melting down—and it may be noted that the claim generally made on this point is that the weld metal from an electrode containing carbon and columbium in the ratio stated above will retain sufficient columbium to prevent sensitization by a second weld crossing or meeting it and not that it will be completely immune from intergranular susceptibility—there still remain the highly heated strips of plate material bordering the deposited metal, which are more susceptible than after normal softening.

It may also be noted that both the weld metal and these highly heated strips bordering it will probably require holding for a longer time at the stabilising temperature than the normally softened material, in order to obtain complete immunity. It has already been pointed out that these parts, when heated at 600° C. or thereabouts, become susceptible appreciably more quickly than if they had been softened at the normal temperature; hence they will be more liable to precipitate intergranular carbide during the slow heating, up to the stabilising temperature, which would inevitably be necessary with a large welded vessel. The duration of the stabilising treatment on a welded vessel should therefore be adequate to ensure removal of the susceptibility produced in these areas while the vessel is being heated to the stabilising temperature. It would obviously be an advantage in such cases if the steels used for both plate and electrode—whether they contain titanium or columbium—should have a carbon content as low as possible; the lower this content, the less effect on susceptibility is likely to be produced by heating to high temperatures.

Bearing all these various points in mind, the problem would appear to be difficult and very probably cases will arise where operating conditions demand that a complete welded vessel shall be most carefully stabilised in order to attain complete freedom from intergranular susceptibility in every part. In other cases, however, the dangers attending the production of some small degree of susceptibility in vessels working in the sensitizing range of temperature would seem to be less serious than might be anticipated. Thus vessels constructed of steel which had satisfied the susceptibility test frequently used in this country—namely, heating for 30 mins. at 650° after normal softening and then immersing for 72 hours in the boiling acid copper sulphate liquor—have been in successful use in the chemical industry in this country, although neither the completed vessels nor the plates from which they were made had been given any stabilising treatment and hence might conceivably develop some degree of susceptibility, particularly at the welded joints, when held for long periods in the sensitizing range of temperature. It may be that in some of these cases the operating temperature was sufficiently high that the corrosive fluid handled was in the form of gas instead of liquid and that such gaseous phases are less corrosive than the liquids from which they are formed,* and that, in addition, the total periods during which the steels were susceptible to intergranular attack were relatively short, immunity being regained fairly quickly as a result of chromium diffusion. It will at least be evident that when the construction of vessels operating at temperatures above about 400° C. are under consideration, many factors have to be taken into consideration in deciding whether or not some form of stabilising treatment is advisable or essential.

Perhaps some day austenitic stainless steels will be produced which are completely free from intergranular susceptibility under any and every condition of treatment. Meanwhile, one must take account of the shortcomings of the available steels. If, however, the latter are used intelligently, there is no reason why intergranular corrosion troubles should occur.

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* For example, acetic acid vapour at 550°/600° C. appears to be considerably less corrosive to "18-8" steels than boiling aqueous solution of acetic acid containing 30% or more of the acid.